BRITISH WELDING JOURNAL

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CONTENTS

Critical Review of Aspects of Cracking in Welded Cr-Ni Steels
By J. C. Borland and R. N. Younger

Effect of Oxygen on CO_2 Welding
By A. A. Smith

Dynamic Behaviour of D.C. Generators
By A. Carrer

Surface Preparation in Cold Pressure Welding
By L. R. Vaidyanath and D. R. Milner

Stresses in Cylindrical Vessels
By R. T. Rose

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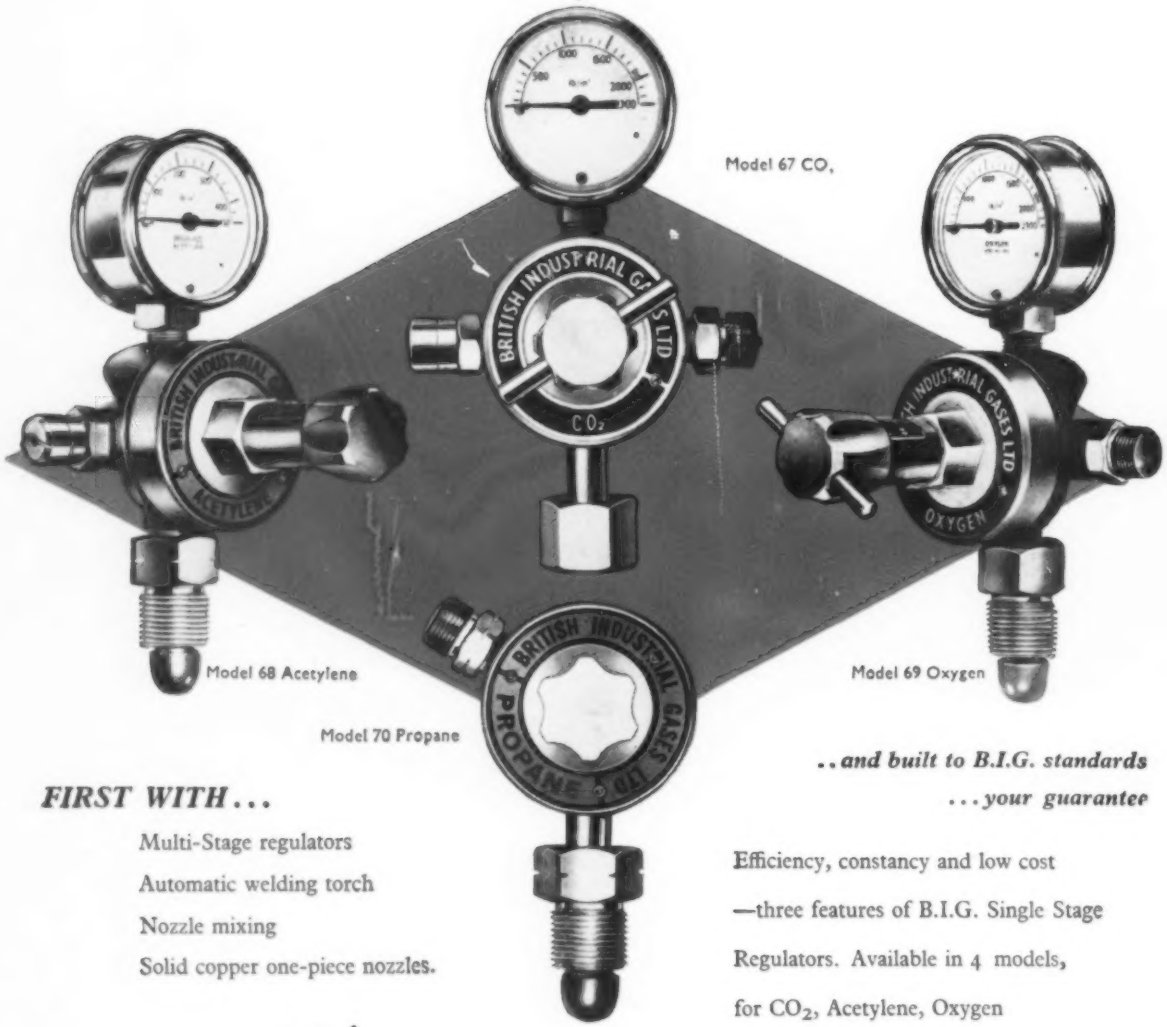

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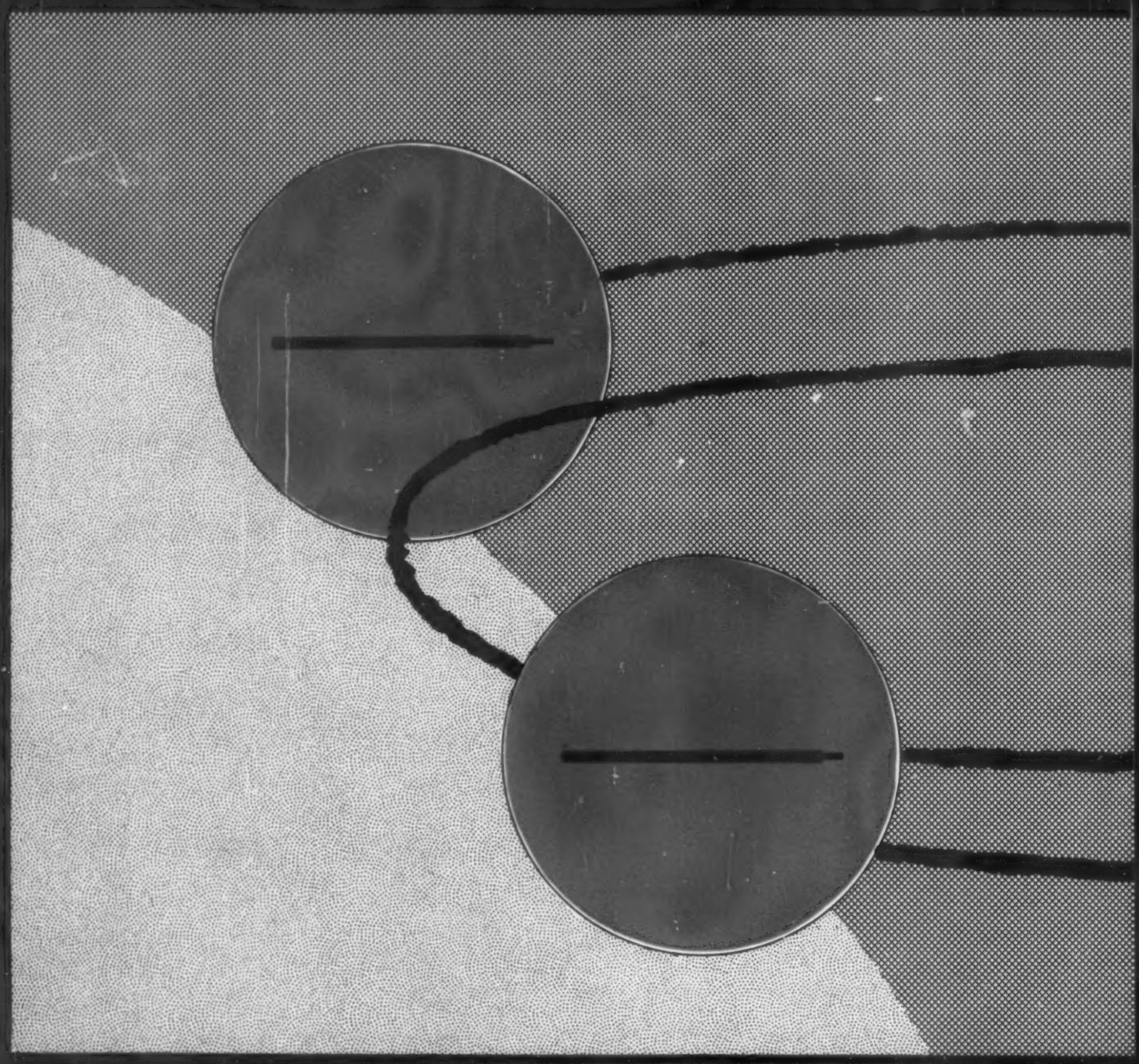

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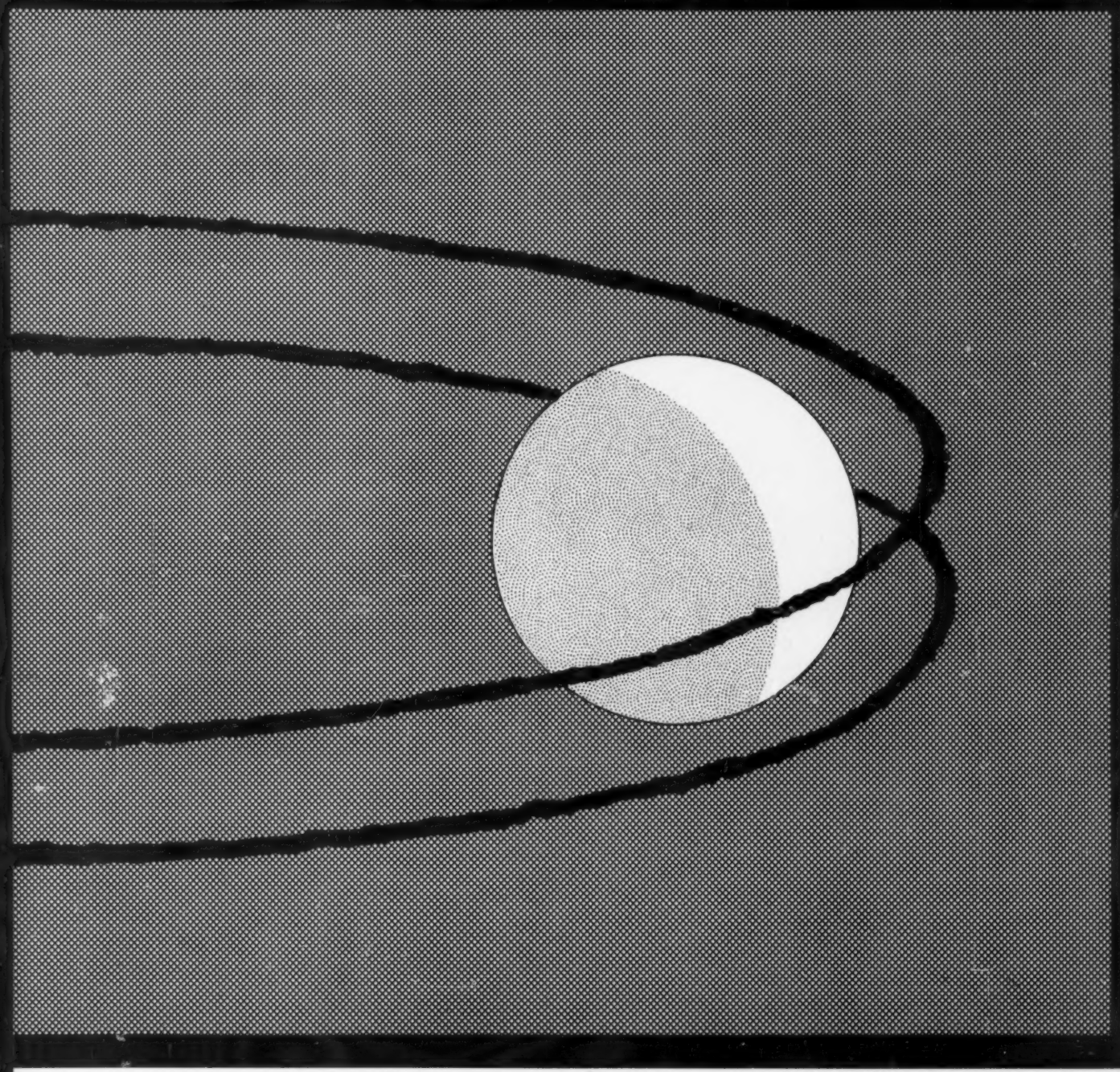

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MIRROSPEED is an outstandingly successful newcomer to the Quasi-Arc range of mild steel electrodes. Welders' appreciation of its long run lengths, the ease with which it can be used, and above all of its exceptional slag detachability, is reflected in the couple of thousand miles it has quickly added to Quasi-Arc's latest million.

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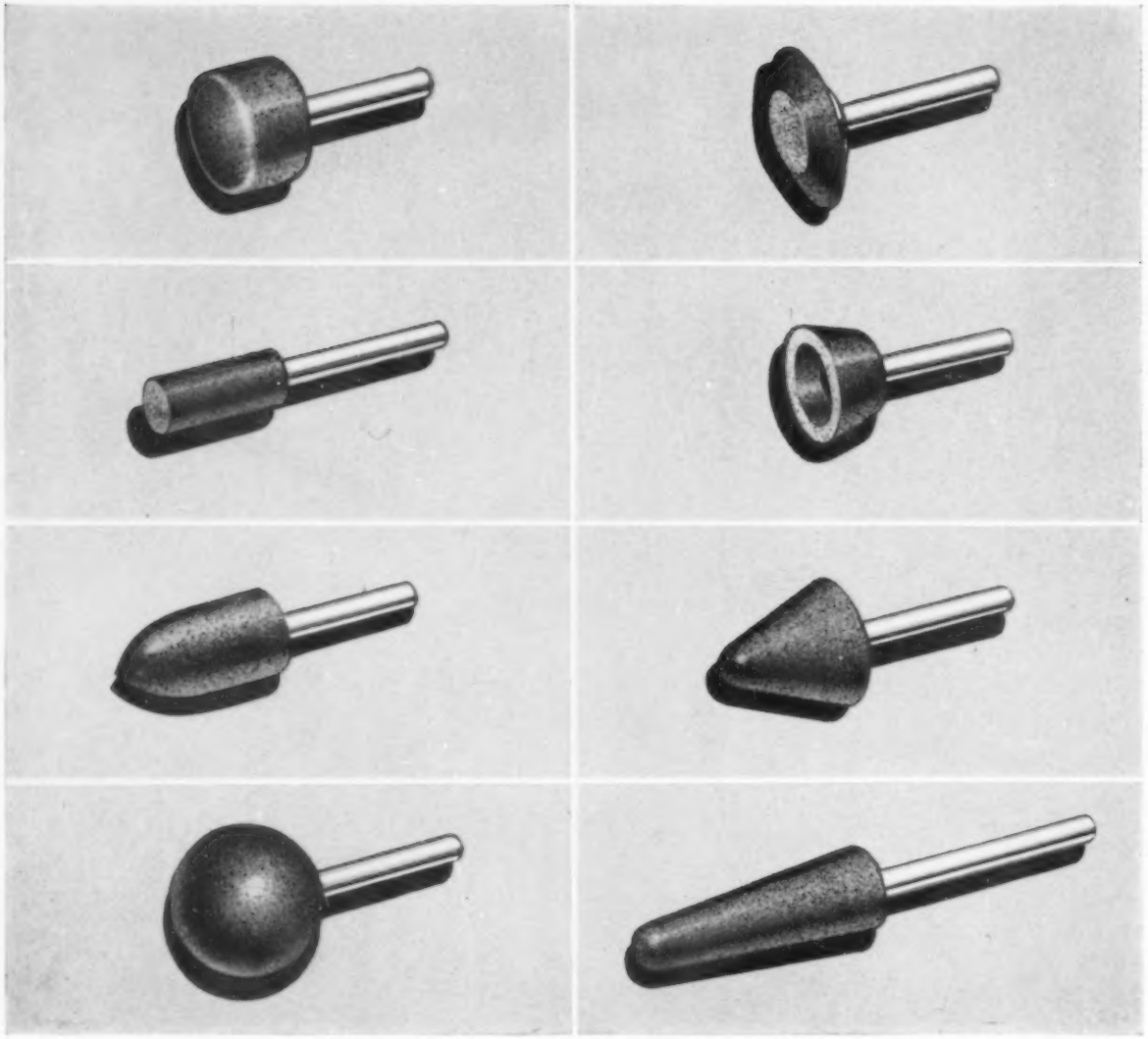

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An Esso Photograph

THE ESSO REFINERY at Milford Haven is a mammoth project. When completed it will be one of the largest oil refineries in Great Britain.

Now past the half-way stage, construction is proceeding rapidly — but carefully. In a complex operation such as this the whole installation could depend on one section, the section on one transfer unit, the transfer unit on a single weld in a single pipe.

Yes, on one single weld. In multi-operational projects like oil refineries (and the same applies to nuclear power stations) the efficient working of the whole plant can depend on the welding of one joint in one pipe.

So at Milford Haven vital welds are examined by radiography. And the inspection engineers depend mainly on 'Kodak' industrial X-ray films to provide the evidence. The high-speed film *Industrex D* is used for most work, and *Crystallex* film is used for those extra-searching inspections where even the most minute flaw must be seen and identified.

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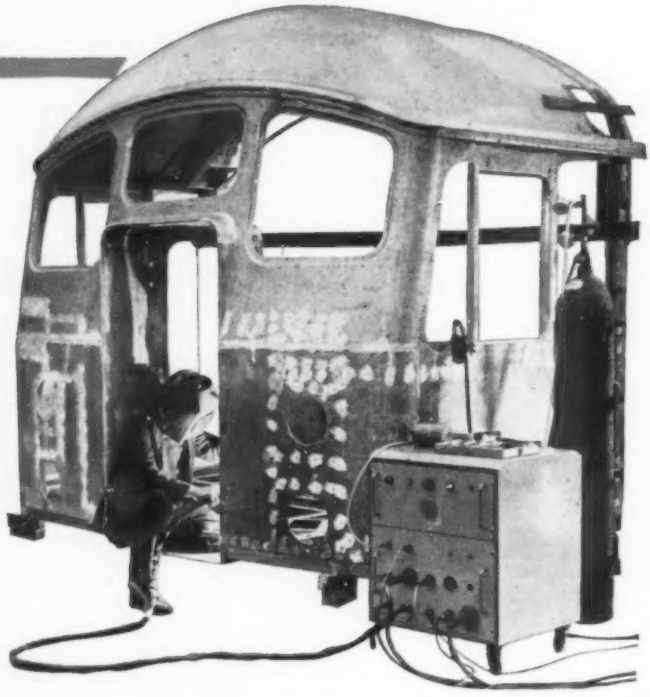

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One of the most modern British Railways locomotives in production today is this 1160 H.P. Diesel electric type, which is being constructed at Derby under the supervision of the Chief Mechanical Engineer (London Midland Region). Saturn-Hivolt argon arc welding equipment and Saturn High purity argon are being used extensively for the important welding of aluminium alloy which is being used to lighten the construction.

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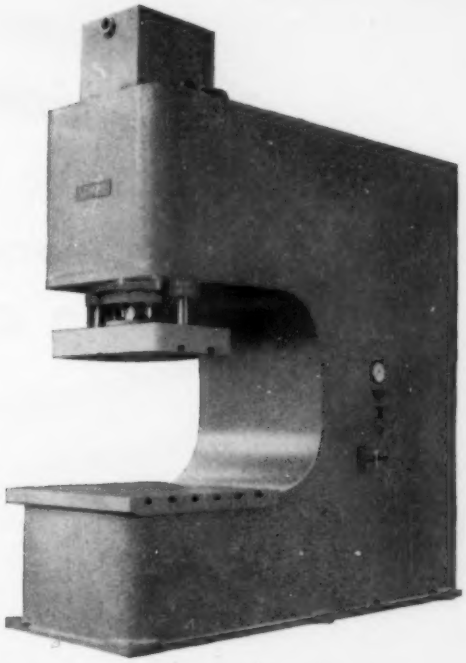
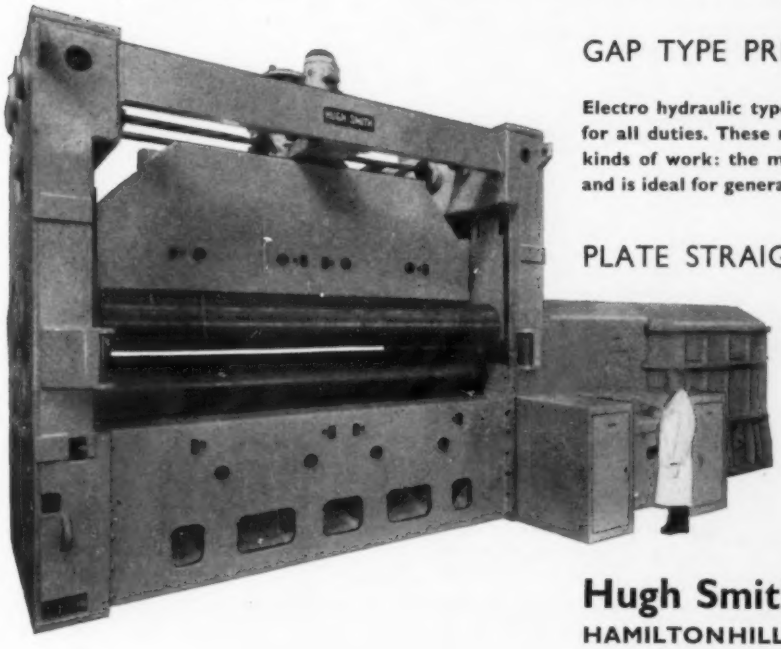


HUGH SMITH
Machine Tools



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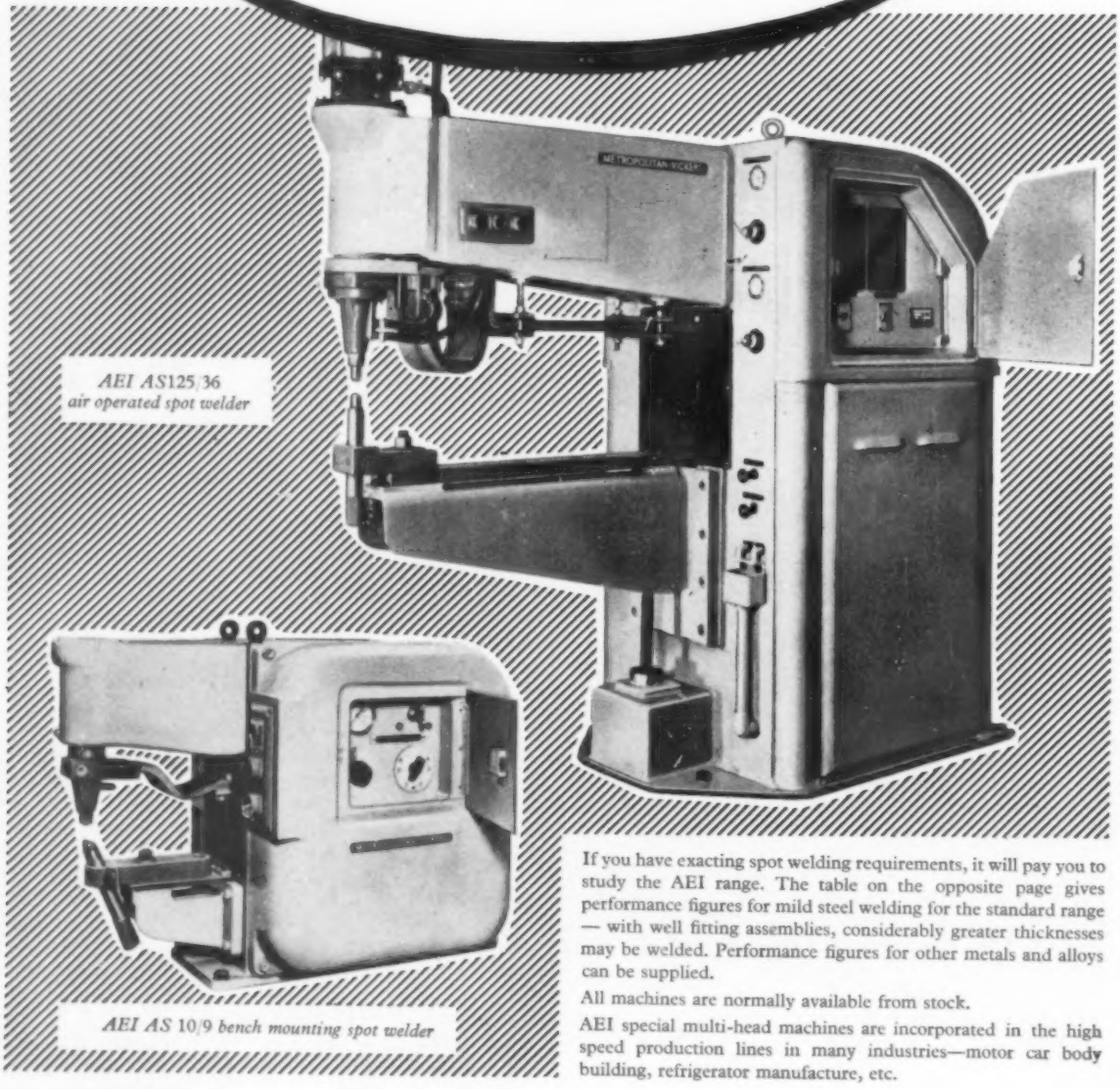

AEI AS125/36
air operated spot welder

AEI AS 10/9 bench mounting spot welder

If you have exacting spot welding requirements, it will pay you to study the AEI range. The table on the opposite page gives performance figures for mild steel welding for the standard range — with well fitting assemblies, considerably greater thicknesses may be welded. Performance figures for other metals and alloys can be supplied.

All machines are normally available from stock.

AEI special multi-head machines are incorporated in the high speed production lines in many industries—motor car body building, refrigerator manufacture, etc.

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STANDARD SPOT WELDERS

Available from stock

	Type	Maximum welding rating*	Throat depth		Maximum welding force†		Maximum capacity
		kVA*	in	mm	lb	kg	
Pedal Operated	M196	4	3 & 6	76 & 152	—	—	2 x 0·036" (0·9 mm) & 2 x 0·028" (0·7 mm) Mild Steel
	M63	16	6 to 18	152 to 456	—	—	2 x 0·11" (2·8 mm) to 2 x 0·078" (2 mm) Mild Steel
	M194	32	12 to 30	304 to 760	—	—	
Air Operated	AS 10/9	16	9	228	250	113	2 x 18 s.w.g. (1·2 mm)
	AS 25/24	35	24	609	600	272	2 x $\frac{1}{8}$" (3·2 mm)
	AS 60/18	90	18	457	1,000	453	2 x $\frac{3}{16}$" (4·8 mm)
	AS 60/24	90	24	609	1,000	453	2 x $\frac{3}{16}$" (4·8 mm)
	AS 75/36	130	36	914	1,000	453	2 x $\frac{3}{16}$" (4·8 mm)
	AS 75/48	130	48	1,218	1,000	453	2 x $\frac{3}{16}$" (4·8 mm)
	AS125/36	215	36	914	2,250	900	2 x $\frac{1}{4}$" (6·3 mm)
Stored Energy (Condenser discharge)	ASC40/36	—	36	914	2,000	906	2 x 14 s.w.g. Aluminium Alloy (2 mm)
Roller (Air operated)	AS100/24R	100	24	609	1,500	680	Dependent upon type of material and application
	AS150/36R	175	36	914	1,500	680	
	AS250/36R	400	36	914	2,250	1,020	

All mild steel capacities based on BS 1140 tests

* Ratings in line with BS 3065:1959

† At 80 lb/sq in (5·6 kg/sq cm) air supply

Send for this booklet

This illustrated booklet gives concise data on these machines and other AEI welding and heating equipment.

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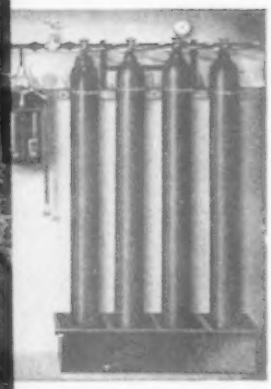
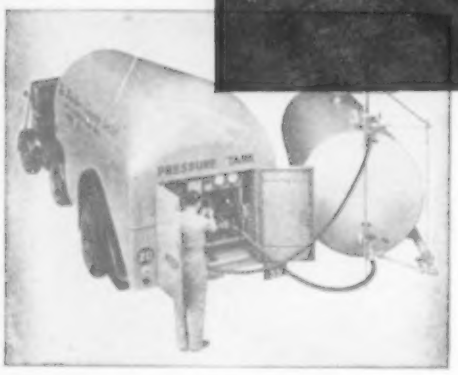
Photograph illustrating the FUSARC CO_2 process by courtesy of Quasi-Arc Limited





The use of low pressure CO_2 gas as an inert shield in the welding of steels is increasing. CO_2 is not only cheap — it enables high quality welds to be made in the shortest time.

Where automatic welding machines have been specially modified to use this technique, the simplicity, ease of slag removal and high speed of welding have been found most impressive. On one differential housing, for example, welding time has been reduced from 105 to 23 seconds.

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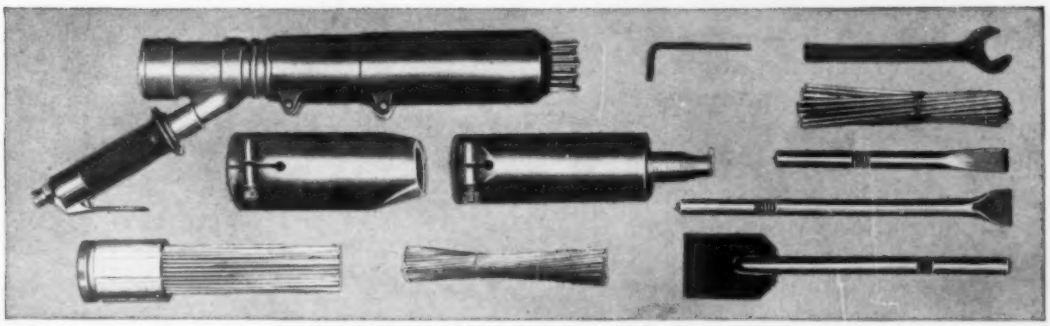

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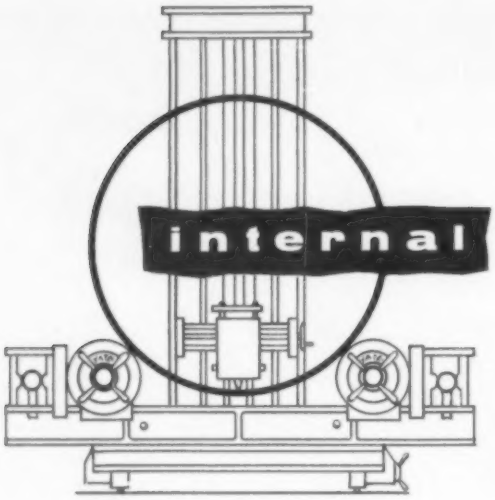

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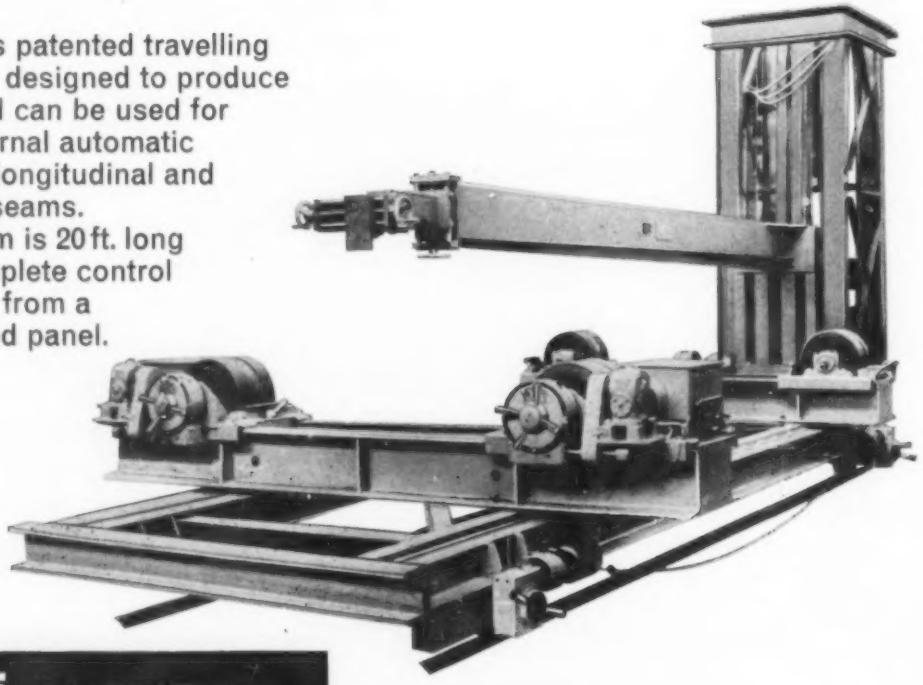

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AEI ELECTRONICS DATA SHEET
Subject IGNITRONS — RATINGS

Welder types

Types	Maximum Demand* (kVA)	Maximum average* anode current (Amps.)	Temperature control
BK 22	450	15	—
BK 24	1200	140	—
BK 24A	1200	140	Integral
BK 24B	1200	140	Clamp on
BK 34	2400	355	—
BK 34A	2400	355	Integral
BK 34B	2400	355	Clamp on
BK 42	600	56	—
BK 42A	600	56	Integral
BK 42B	600	56	Clamp on
BK 66	300	22·4	—

* Ratings are for welder control service and refer to two valves in inverse parallel at any voltage from 250-600v. r.m.s.

Rectifier types

Type	Maximum peak voltage (Kilo volts)	Maximum average current at peak voltage (Amps.)
BK 44	2·1	75
BK 46	2·1	150
BK 56 †	20·1	150

† Tentative ratings.

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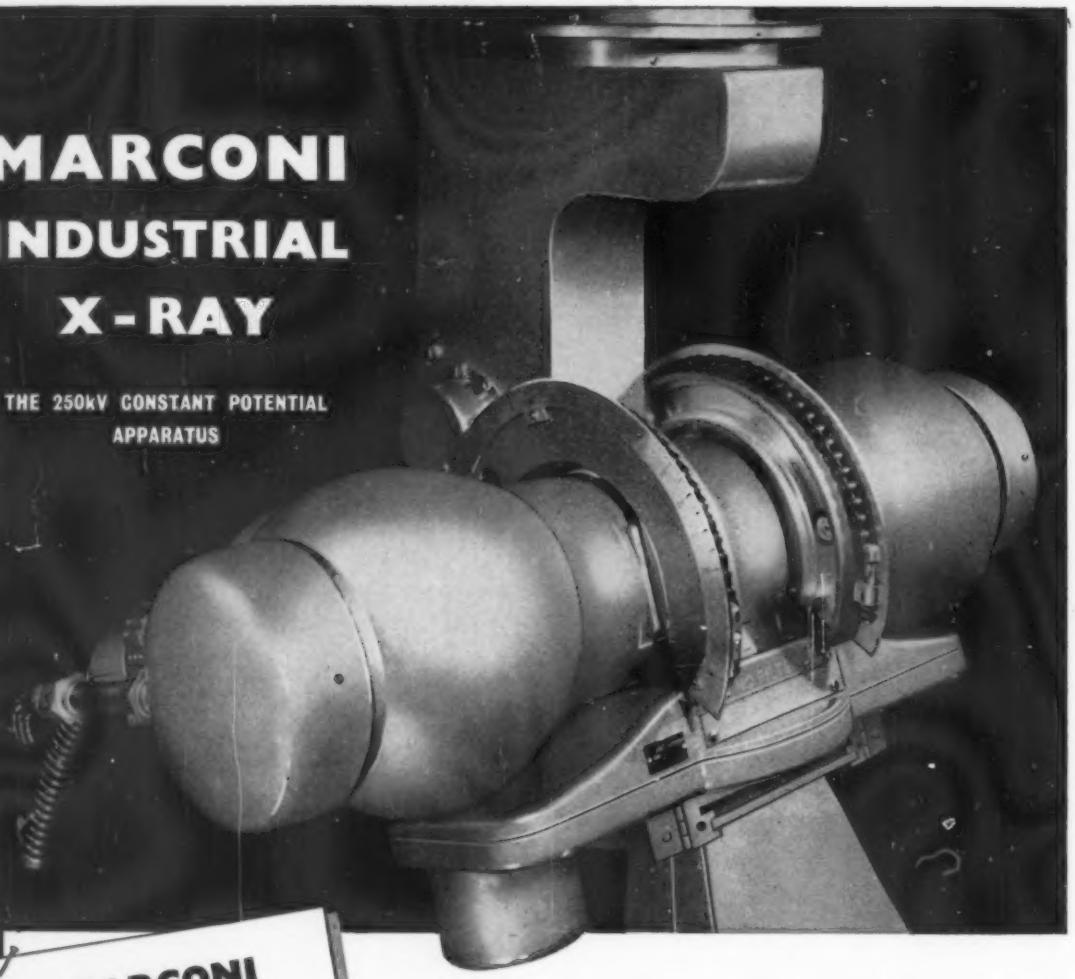
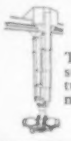


MARCONI for X-Ray

SOME MOUNTINGS FOR THE 250-kV TUBEHEAD *



Telescopic-suspension tubehead mounting

Double-column gantry tubestand

Dual-travel tubestand

* *Let Marconi engineers help you in your choice of mountings. In addition to those shown, other tubehead mountings are available to suit specialized requirements. Full details gladly sent on request.*

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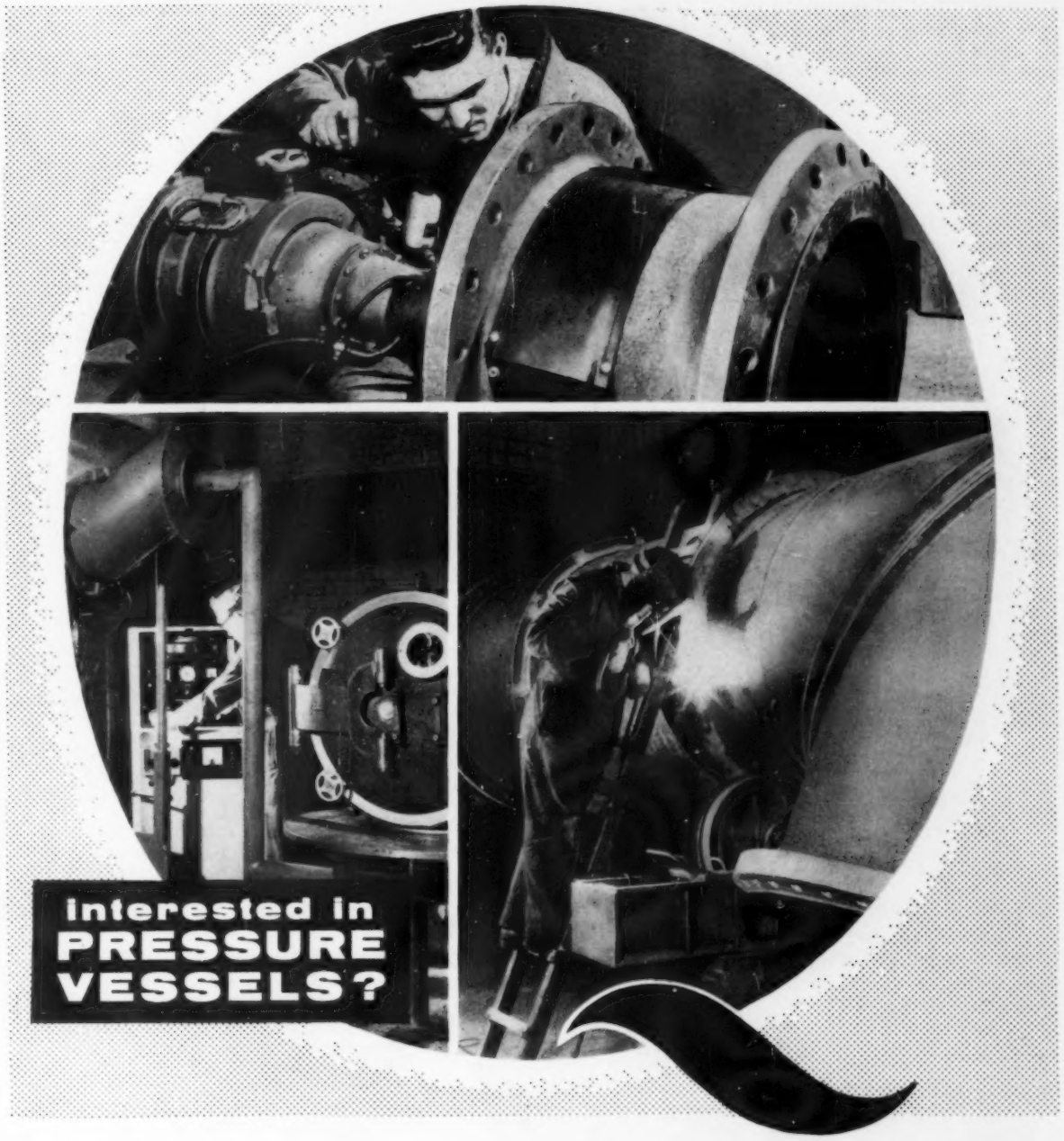
IX22



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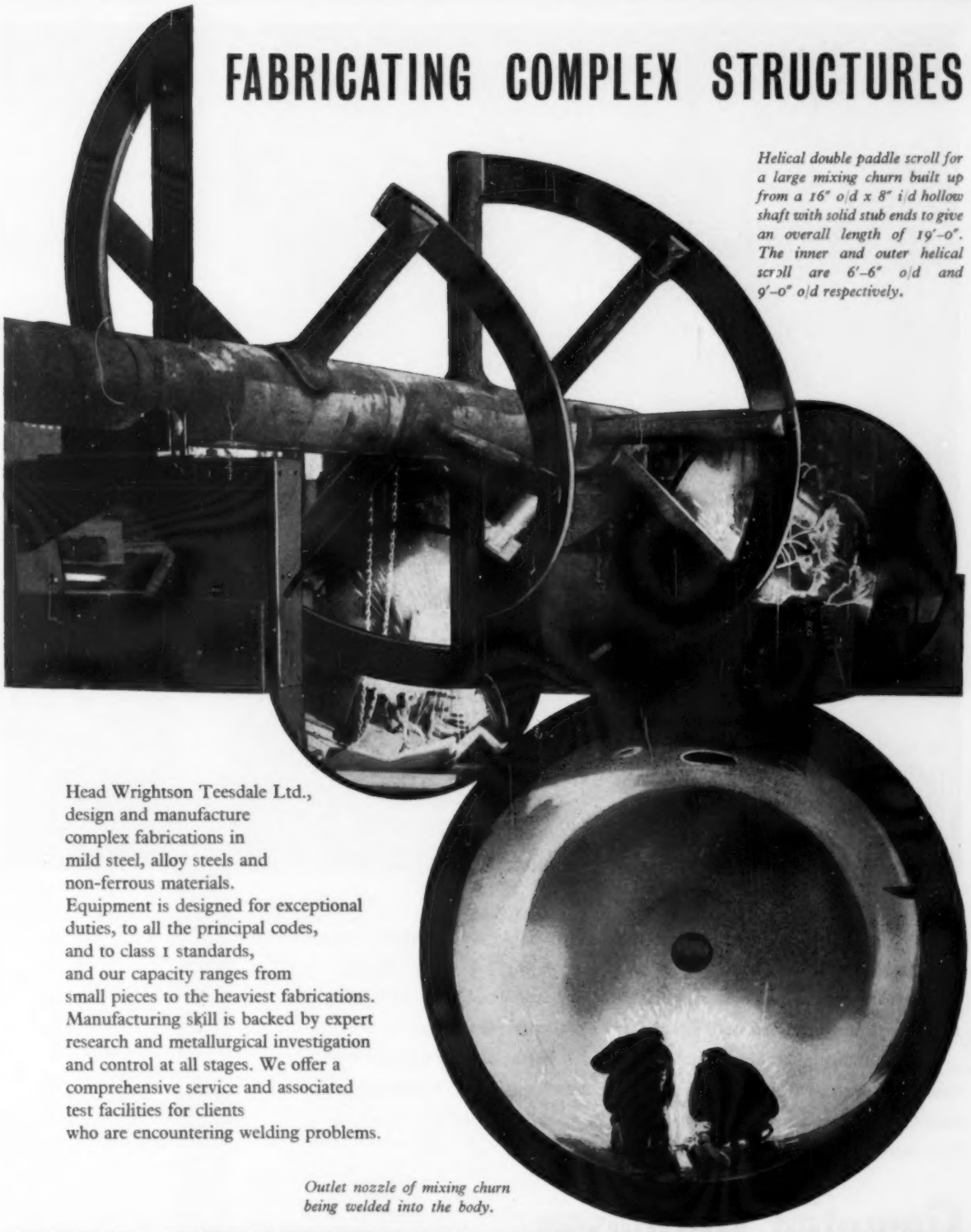

Helical double paddle scroll for a large mixing churn built up from a 16" o/d x 8" i/d hollow shaft with solid stub ends to give an overall length of 19'–0". The inner and outer helical scroll are 6'–6" o/d and 9'–0" o/d respectively.

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Outlet nozzle of mixing churn being welded into the body.

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P.1482

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and simplest way
of fixing
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to this? . . .

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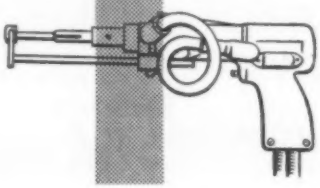
STUD WELDING



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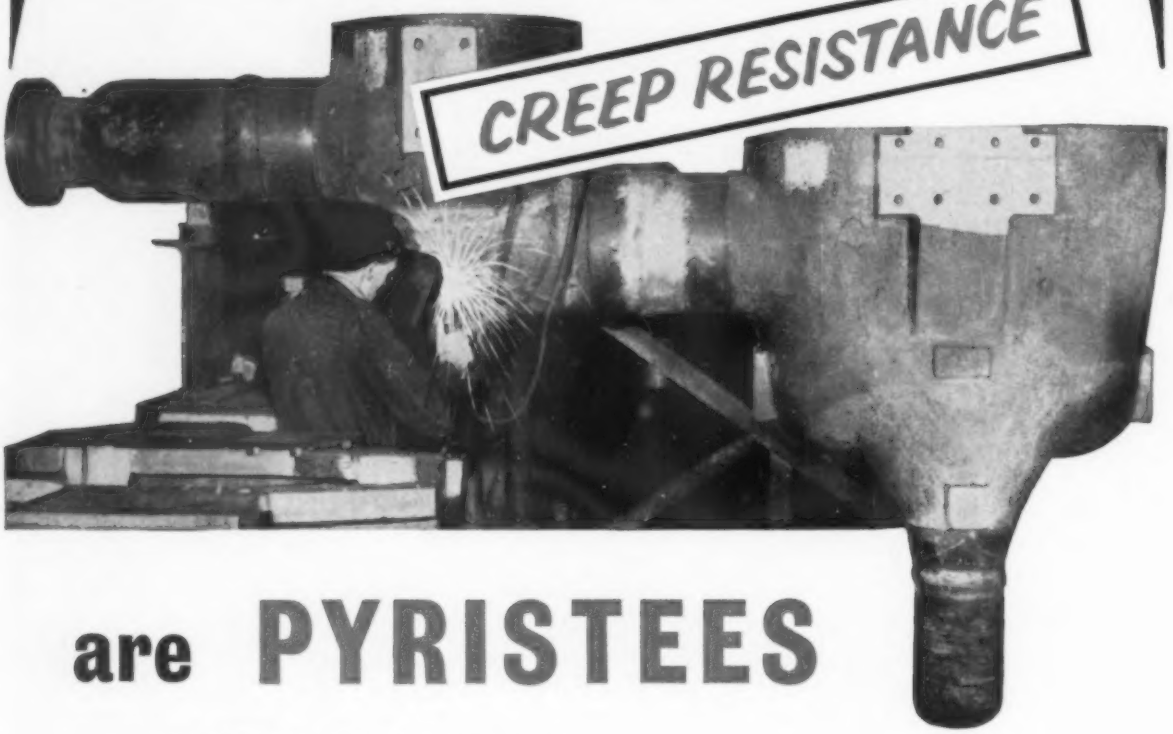

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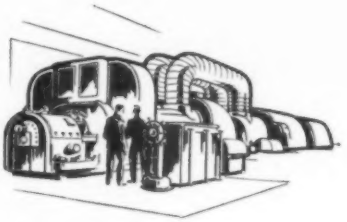

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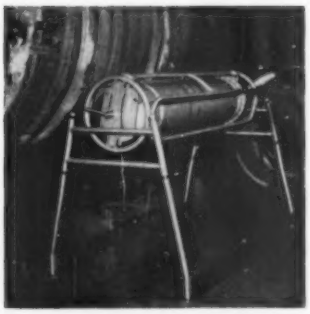

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CONTENTS

INSTITUTE OF WELDING

1 **Significance of Surface Preparation in Cold Pressure Welding**
By L. R. Vaidyanath and D. R. Milner

INTERNATIONAL INSTITUTE OF WELDING

6 **Dynamic Behaviour of D.C. Generators for Arc Welding**
By A. Carrer (Commission II)

BRITISH WELDING RESEARCH ASSOCIATION

15 **Effect of Oxygen Additions to the CO_2-Shielded Metal Arc**
By A. A. Smith

19 **Stresses in Cylindrical Vessels Due to Local Heating Stress Relief of Circumferential Welds**
By R. T. Rose

22 **Aspects of Cracking in Welded Cr-Ni Austenitic Steels: A Survey of the Literature**
By J. C. Borland and R. N. Younger

60 NEWS AND ANNOUNCEMENTS

65 CURRENT LITERATURE

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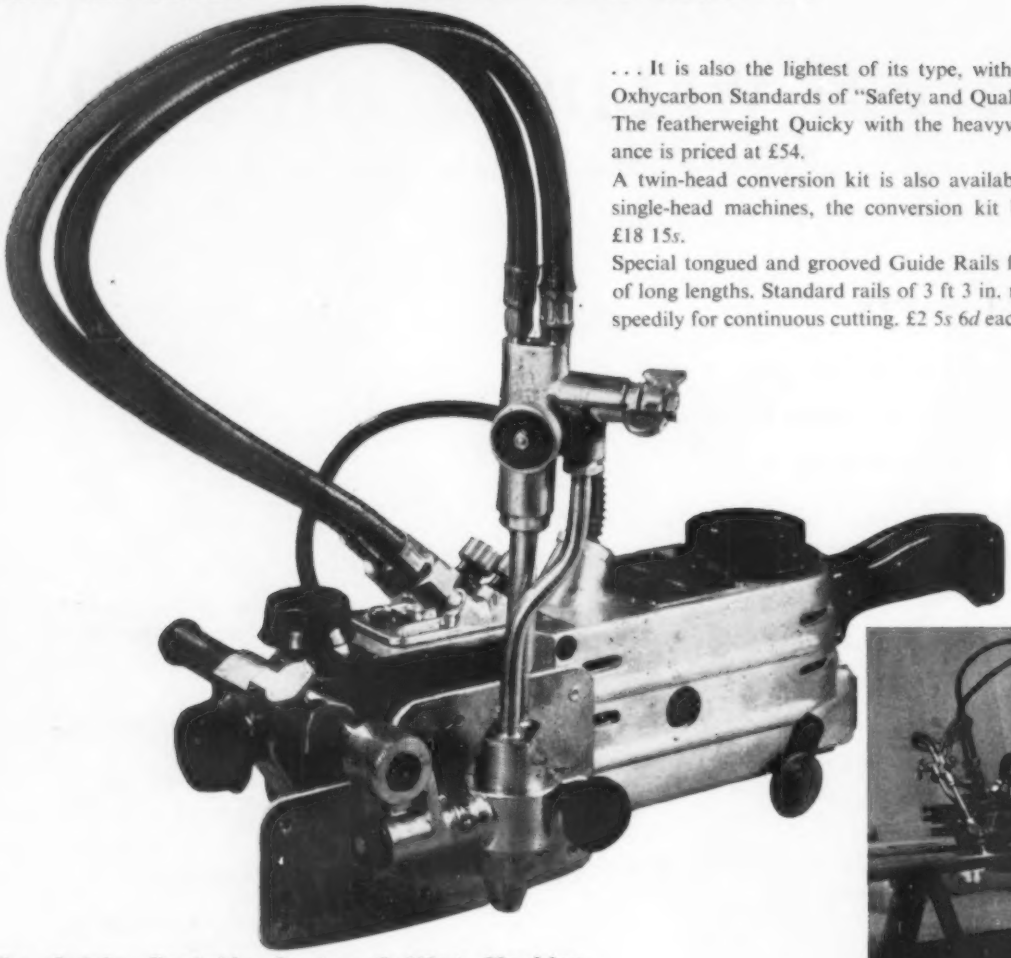

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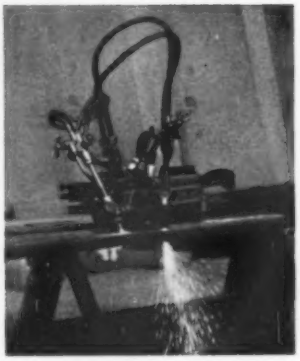

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Welding
TECHNOLOGY

Significance of Surface Preparation in Cold Pressure Welding

By L. R. Vaidyanath and D. R. Milner

The bond strength characteristics of aluminium and copper have been determined at room temperature after various types of surface treatment. All surfaces can be welded if they are first 'baked-out' by heating at several hundred degrees centigrade. The highest bond strengths are obtained with scratch-brushed surfaces; anodized aluminium surfaces also have good bonding properties after baking out, but in the case of copper a thick oxide film fractures by shear and spreads along the interface, thus reducing the area of potential bonding.

It is concluded that scratch-brushing is a particularly effective form of surface preparation for cold welding because:

(1) It removes surface contaminants and allows welding to be carried out before they have time to be readsorbed.

(2) It forms hard rough layers on the two surfaces to be welded which 'adhere' to each other during their progress through the roll gap and thereby break up as one so exposing the maximum area of virgin metal for potential bonding.

Introduction

THIS work is part of a programme of research into the mechanism of pressure welding by rolling, aimed at developing a model which will explain the relative importance of all significant variables and, where possible, forming a basis for approximate quantitative predictions. The results of the first phase of the programme have been reported in a paper[1] which dealt with the mechanism of cold pressure welding as it is operative during roll bonding. From this research the two following conclusions emerged which it is intended to extend and modify as required by the results of the present and future work, so as to establish the underlying principles of the roll bonding process:

(i) The maximum bond strength that can be achieved at room temperature with scratch-brushed surfaces is determined by the fact that scratch-brushing produces a heavily worked layer, on top of which the normal oxide film forms: at an early stage of their progress through the roll gap the layers on the two members of the composite come together and thereafter deform as one, and being completely brittle, the area available for potential bonding is the virgin metal produced by the increase in area of the interface.

This conclusion was derived from experimental observations on aluminium and copper, but a theory of the development of bond strength with increasing degrees of deformation, based on this data, was found to apply also to tin and lead in which the scratch-brushed layers recrystallize at room temperature.

(ii) Whether or not the virgin metal areas weld is determined by several parameters; *viz.*, surface preparation, exposure of the surfaces to the atmosphere before welding, pressure, and the time and temperature of welding. On the evidence available it was not possible to explain the significance of these factors.

The present paper describes the continuation of this work, in which an attempt has been made to elucidate

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the role of surface preparation in the mechanism of cold pressure welding: this in turn has lead to the study of the effect of layers of contaminants adsorbed on the surfaces to be welded, and the effect of surface roughness and oxide film thickness. The significance of the time and temperature of welding are being examined in an investigation by Nicholas, which will be reported at a later date; additional work is also in progress on the problem of bonding between dissimilar metals.

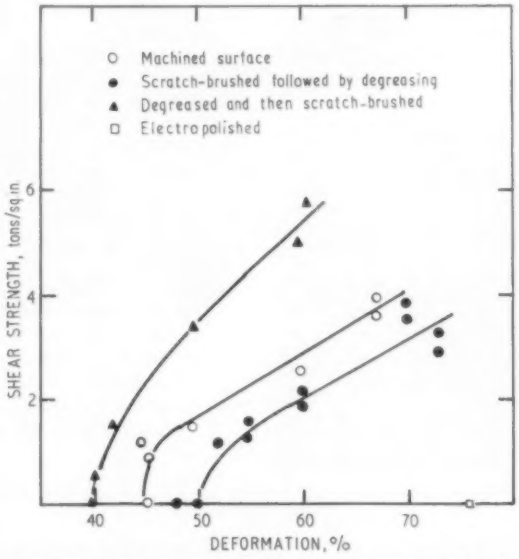
Whenever metals are to be pressure welded at room temperature the question of preparation arises, for it is invariably found that welding will not take place without some sort of surface treatment. The importance attached to the technique of surface preparation has led to a number of investigations and hypotheses, but as yet no conclusive theory has been evolved. For cold pressure welding it has been found that degreasing followed by scratch-brushing immediately before welding gives the highest bond strengths. As some preliminary experiments showed this to be equally true for roll bonding, degreasing in trichlorethylene vapour, followed by scratch-brushing and rolling within two minutes, was adopted as the standard technique for the work described in the previous paper. It is often stated in technological practice that scratch-brushing removes scale and oxide films; although this may be true it cannot produce a surface which is free from oxide, for the scratch-brushing operation heats the surface layers (probably to the melting point[2]), so that a further oxide film will be rapidly reformed before welding can be carried out. Parkes[3] has suggested that surface roughness is important, in that rough surfaces will only make contact over small areas, therefore giving low bond strengths. This will be true only for the infrequent case of low loads and little deformation as used by Parkes, and is in direct opposition to the general observations regarding the superiority of scratch-brushing. Durst[4] has postulated that the prime effect of scratch-brushing is the removal of adsorbed contaminant surface layers. This finds support from the first phase of this work[1] when it was shown that if scratch-brushed surfaces were exposed to the atmosphere before welding, for periods exceeding about two minutes, the bond strength progressively decreased owing to the adsorbtion of contaminants, probably principally water vapour. Normal surfaces, or chemically treated surfaces, as well as surfaces mechanically cleaned by a prolonged operation, would be expected to possess adsorbed contaminant layers, so that the starting point of the investigation was to remove these layers by 'baking out' and to determine the subsequent effect on the welding characteristics.

Experimental

Using the experimental technique and types of specimen described in the previous paper, two-ply composites were roll bonded and the weld strengths were determined by isolating a small area of the joint and testing it in shear. The results are presented either in terms of the development of the bond strength at the centre of the composite (initially measuring 4×4 in.) as a function of deformation, or as the strength across the width of the composite at constant deformation, depending on which is the most appropriate. It was not possible to extend the researches to the wide range of metals considered in the earlier investigation, since adsorbed contaminant had to be baked off at high temperatures (so that tin and lead, for example, could not be used); the bulk of the investigation was therefore carried out with aluminium, and the conclusions were examined for copper in a limited series of experiments.

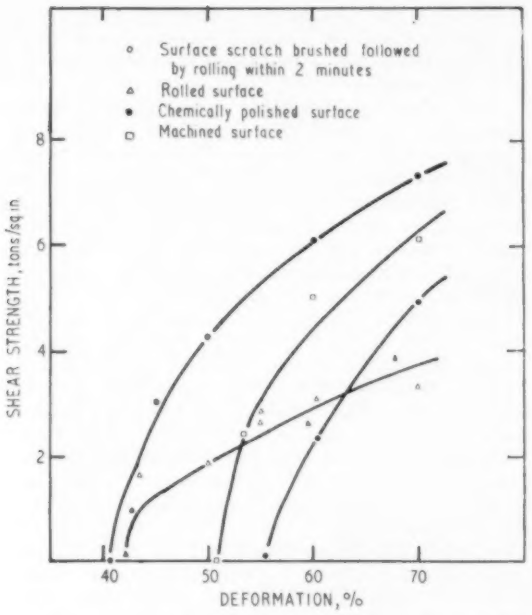
Influence of adsorbed contaminants

To determine the relative importance of adsorbed layers of contaminants, specimens of annealed aluminium which had undergone various surface treatments, or no surface treatment, were welded and the results were compared with those obtained from similar specimens which had been heated in air to remove adsorbed layers and then cooled in a desiccator to prevent subsequent readsorption of moisture. With the surfaces which had not been heated it was found (Fig. 1), as expected, that degreasing followed by scratch-brushing gave the best bonding properties;

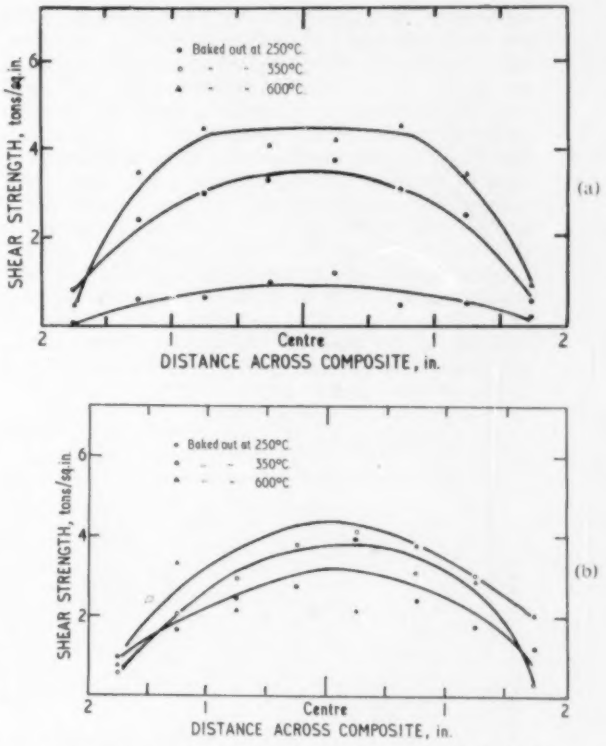


1—*Effect of different types of surface preparation on the bonding of aluminium composites*

when the procedure was reversed, with scratch-brushing followed by degreasing, the threshold deformation for bonding was higher, and lower bond strengths were obtained. Machined surfaces gave intermediate bond strengths, while electropolished surfaces, and the 'as-received' commercial surfaces, either with or without a degreasing treatment did not weld even with an 80% deformation. By comparison, it was found that after heating to 500°C. in air and cooling in a desiccator, chemically polished surfaces and even the untreated commercial type of surface could be welded, although the bond strengths were not as high as those obtained with scratch-brushed surfaces (Fig. 2). This marked improvement in bonding brought about by heating followed by cooling in a desiccator thus suggests that a major action of good



2—*Bond strength characteristics of various types of aluminium surfaces which had been heated in air at 500°C. and cooled in a desiccator before welding, with scratch-brushed surfaces included for comparison*



3—*Improvement in bond strength of composites which had undergone 60% deformation brought about by 'baking-out' at various temperatures*: (a) *In air*, (b) in vacuo

surface preparation is the removal of layers of adsorbed surface contaminant.

Since the temperature at which adsorbed contaminants are removed is of some practical importance, a further series of tests was undertaken to relate the baking-out temperature of the as-received commercial surface to the improvement in bond strengths, and to evaluate the relative advantages of heating aluminium in air or a vacuum (0·005 mm mercury column). In this case all composites were given a 60% reduction and the variation in bond strength was determined across the width. No bonding could be obtained by heating in air at temperatures below 250°C. or below 180°C. in vacuum; thereafter the bond strength increased progressively with 'baking-out' temperature and there was relatively little difference between air and vacuum heating (Fig. 3). However, even baking-out at temperatures as high as 600°C. did not give such strong bonds as scratch-brushing; the maximum central strength was about 4·5 tons/sq.in. compared with 6 tons/sq.in. for similar specimens which were roll bonded after scratch-brushing (from Fig. 16, ref. 1). It would appear therefore that although adsorbed contaminants account for much of the difference in bonding characteristics between scratch-brushed and other surfaces there must also be another significant factor, since heating at 600°C. should have completely removed all adsorbed material.

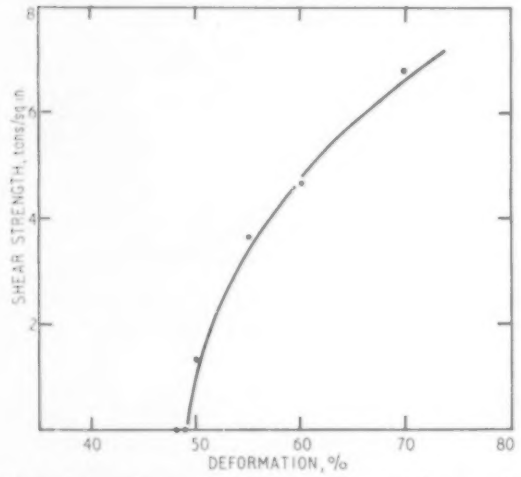
Experiments with thick oxide films

While the bond strength measurements given in Figs. 1–3 suggest that another factor besides the removal of adsorbed layers is important in achieving good bonding, they give no suggestion as to what this other factor can be and the alternative source of information, *i.e.*, metallographic observation of interfaces, does not lead to very positive information where only thin oxide films are present. Attention was therefore turned to the welding of anodized surfaces where the fragmented oxide could be studied with an optical microscope.

Some workers, and in particular Tylecote,[5] have suggested that weldability decreases as the thickness of the oxide film increases, but against this argument Donelan[2] has recently shown that aluminium surfaces possessing a thick anodized film have in fact good bonding properties. In the present work, when roll bonding of anodized aluminium was first attempted, bonding was impossible, even with an 80% deformation, but if surface contaminants were baked off at 500°C. and the composite cooled in a desiccator, good bonding properties were obtained. When the bond strengths, shown as a function of deformation for baked-out anodized surfaces (Fig. 4), are compared with similar results for other surface preparation (Fig. 2), it is seen that an anodized surface has bonding properties superior to all other treatments tried except scratch-brushing. Nevertheless the difference between the bond strengths of scratch-brushed surfaces and those of anodized surfaces represented significant differences in the area of interface welded; the reason for this was therefore sought by microscopic examination. Figure 5 shows a typical interface with anodized

surfaces for 60% deformation in which the break up of the oxide films can be clearly seen. For the most part the two oxide films have fragmented together, although a sufficient proportion of oxide has broken into single-layer particles to exert an effect on the bond strength. Detailed measurements showed that 33% of the interface was occupied by the double layer of oxide, 23% by a single layer, and the remaining 44% apparently bonded. By comparison,[1] two scratch-brushed layers in aluminium (and copper) always break up as one to give a total length of discontinuity equal to the original length of the specimen, which in this case would mean a 60% bond area compared with the 44% obtained with the anodized surfaces. When the strength of the bond in the anodized specimen is calculated from the formula for bond strength derived in the previous paper, *viz.*:

$$\frac{\text{Ultimate shear strength of the weld}}{\text{Ultimate shear strength of solid metal}} = A(2-A) \quad \ldots\ldots (1)$$



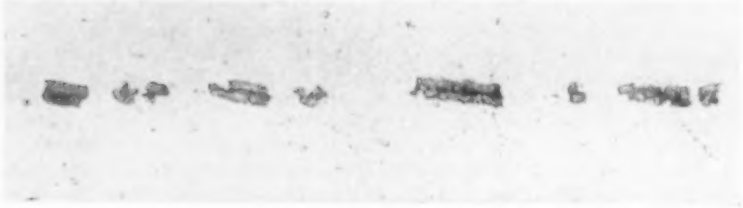
4—*Bond strength versus deformation characteristic curve for anodized aluminium baked out at 500°C. and cooled in a desiccator before welding*

where A is the proportion of interface apparently bonded, a value of 4·5 tons/sq.in. is found, which agrees well with the experimentally determined value of 4·75 tons/sq.in.

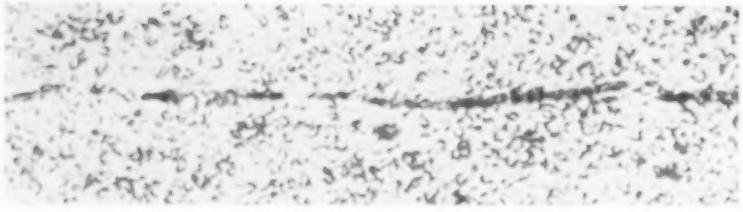
Thus it can be concluded that pressure welds in anodized aluminium do not give the high bond strengths obtained with scratch-brushed surfaces because two anodized surfaces do not break up as one.

It is now necessary to decide whether this concept, derived from anodized surfaces, leads to an explanation for the low bond strengths obtained with other types of surface preparation in which the oxide film thickness would be considerably less; *e.g.*, of the order of 1000 Å for the as-received rolled sheet. For this to be so, it would infer that the oxide films had broken up even more independently to give a lower area of bonding. This would be difficult to prove directly, for the determination of the distribution of the oxide particles would require a degree of resolution far greater than that obtainable with an optical microscope. However, on etching heavily in 2% HF, such interfaces showed a structure consisting of a series of discontinuities smaller than those occurring with scratch-brushed or anodized surfaces (Fig. 6), and measurements made on a specimen which had undergone 60% deformation showed an apparent bond, after etching, equal to 21% of the interface. This is compatible with the hypothesis that thin oxide films on a normal metal substrate tend to break up independently of each other.

The experiments so far reported were all carried out using aluminium, and it was therefore considered to be desirable that the main conclusions should be verified with another metal, for which copper was chosen. Again it was found that scratch-brushing was the most satisfactory preparation for cold roll bonding but that, if the specimens were first baked-out and then cooled in a desiccator, other types of surfaces could also be welded, albeit with lower bond strengths. Particularly significant were the results obtained with thick oxide films, which in this case were formed on commercial rolled surfaces which were first heated *in vacuo* at 500°C. to remove contaminant layers,



5—*Interface of a weld between two anodized aluminium surfaces, deformation 60%* ×575

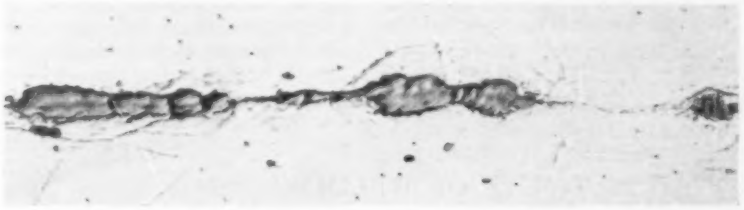


6—*Appearance of the interface of a weld between rolled aluminium surfaces after heavy etching 2% HF, deformation 60%* ×1500

exposed to the atmosphere for two minutes at the same temperature to form an oxide film $7 \cdot 10^{-4}$ cm thick, and then cooled in a desiccator. It was found that the threshold deformation was increased from 45% for scratch-brushed material to about 60% for the heavily oxidized metal, but owing to the difficulty of obtaining higher deformation in one pass with the rolling mill available it was not possible to determine a complete bond-strength/deformation curve, the results being limited to two deformations:

Deformation, %	*Bond Strength, tons/sq.in.*
60	1·5
63	1·7

The increase in the threshold deformation, as compared with the scratch-brushed surface, was greater for copper than for aluminium because of the oxide break up at the weld interface as shown in Fig. 7. Whereas with anodized aluminium the oxide fragmented into discrete rectangular blocks, with copper the oxide has sheared and fractured in an irregular manner, so that the total length of the oxide is greater than the original length of the specimen by a factor of 2·0, with only 18% of the interface available for bonding, compared with the 60% which would be available with the scratch-brushed surface.



7—*Interface of a weld between heavily oxidized copper surfaces* ×575

Discussion

From the work described it would appear that as a technique of surface preparation scratch-brushing serves two functions: Firstly, it removes adsorbed layers of contaminants and allows welding to be carried out before they have time to reform; secondly, it forms a heavily worked layer with properties causing it to 'adhere' to another scratch-brushed surface so that the two layers break up as one. No detailed mechanism has been presented for either of these functions but it is possible to discuss them in general terms to show that qualitatively they are not unreasonable, and that they can be fitted into the model of pressure welding by rolling which was proposed in the previous paper.

The significance of adsorbed layers is not difficult to appreciate. Thus the 'maximum-weld-strength' model previously postulated is based on the concept that scratch-brushing forms a hard and brittle layer on top of which the surface oxide film forms. As the composite is extended by rolling, the brittle layers on the two welding surfaces break up together to expose virgin metal areas for bonding. It is assumed that the brittle layers themselves will not join because of the higher temperature required for the bonding of the oxide films. If adsorbed contaminants are present, then by their very nature they are attached to the oxide by weak Van der Waals' forces, since an oxidized surface has only a low free energy. When the brittle layer becomes fragmented and exposes the adjacent areas of virgin metal, the adsorbed contaminants will have a strong preference for attachment to the high-energy clean-metal areas and thus inhibit the welding of these regions. This transposition of the contaminant must occur very rapidly, for each element of the composite is in the roll gap only for a period of the order of 0·05 sec.

The problem of analysing the factors which determine whether two surface layers break up as one or independently is more complex. It is clearly part of the overall problem of the break up and disposition of the surface oxide layers, which in the present example occurs as the composite is reduced in the roll gap. No model has yet been developed which deals with this problem, and it appears to be of somewhat secondary importance when considering the macroscopic parameters of significance in pressure welding, although some evidence is accumulating which suggests that it may be very important when considering a detailed mechanism of solid-phase bonding. Only a few general remarks can be proffered here.

As the two members of the composite approach the roll gap the surfaces will come into contact, and if the two surface layers break up as one they remain in contact, and the virgin metal areas created are formed opposite each other. If, however, they are to break up independently then in some areas the oxide opposite sections of one layer must be replaced by virgin metal, and this infers that the fragments of the disrupted surfaces must be able to undergo small movements relative to each other. Whether or not this can come about will depend on the coefficient of friction, which will be higher for rough than for smooth surfaces.

This argument is in agreement with the general trend of the bond strength results given in Figs. 2 and 4. which show that at high deformation, *e.g.*, 70%, bond strengths increase from the relatively smooth rolled and chemically polished surfaces, through machined and anodized surfaces, and are a maximum with the rough scratch-brushed surface. (It is necessary to consider a high deformation for at low deformations all the virgin metal area available is not necessarily welded, see ref. 1). It is also interesting to note that the lowest bond strength at 70% deformation, *i.e.*, 3·8 tons/sq.in. obtained with the rolled surface, is

exactly the strength that would be expected if the two oxide films broke up completely independently of each other (*i.e.*, 49% bond area, determined from the strength by equation (1)), whereas the highest bond strength, obtained with scratch-brushed surfaces, corresponds to the two surfaces breaking up as one, while the strengths given by other types of surface are intermediate between these extremes.

The anodized surfaces give higher bond strengths than would be expected from considerations of roughness alone thus suggesting that oxide film thickness may also be significant; the interpretation of this fact would necessitate a knowledge of the forces which cause the small amount of relative movement, but at present little is known of the origin of these forces.

Conclusions

Scratch-brushing is a particularly effective form of surface preparation for cold roll bonding because:

(i) It removes layers of adsorbed contaminants and allows welding to be carried out before they have time to reform

(ii) It forms hard rough layers on the two surfaces to be welded which 'adhere' to each other during their progress through the roll gap and thereby break up as one, so exposing the maximum of area of virgin metal for potential bonding.

Acknowledgments

This work is part of a programme of research into some aspects of the joining of metals which is being carried out in the Department of Industrial Metallurgy at the University of Birmingham. The authors wish to thank Professor E. C. Rollason, Head of the Department, for his interest and encouragement throughout the course of this work, and their colleague M. G. Nicholas for stimulating discussions.

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INTERNATIONAL INSTITUTE OF WELDING

Dynamic Behaviour of d.c. Generators for Arc Welding

PROPOSAL FOR A DEFINITION OF SUITABILITY FOR WELDING

By A. Carrer (Italy)

Oscillographic studies of the behaviour of several d.c. welding generators have made it possible to define the "suitability for welding" of a generator in terms of a coefficient τ, *or less precisely by a coefficient* ρ. *These values are practically independent of the skill of the operator and of the type of basic electrode used.*

The suitability of a generator for welding is considered to be satisfactory if τ *is greater than 0·6 or* ρ *is greater than 0·7.*

General

ELECTRO-MAGNETIC generators are commonly designed to satisfy a number of particular duties and, in general, it is relatively easy to formulate tests which are suitable for determining whether the machine conforms with the specified duties.

In addition to compliance with general clauses and tests for temperature rise, efficiency, insulation resistance, etc., carried out at machine settings corresponding to nominal and maximum currents, arc welding equipment in general, and welding generators in particular, must also provide a specific quality of performance known as "suitability for welding".

The suitability for welding of a given generator must be understood as the ease with which an operator of average skill is able to strike and maintain a stable arc when using covered electrodes of normal quality in the welding positions for which they are suited.

Suitability for welding cannot be judged by the characteristics and quality of the weld deposit, which depend essentially on the type of electrode used.

Report IIW/IIS–11–59 (ex doc. II–31–58E, IIB–40–58F) of Sub-Commission B "Testing and measurement of arc welding equipment" of Commission II "Arc welding".

The research was carried out by the Italian Institute of Welding in collaboration with the Electrotechnical National Institute "G. Ferraris" of Turin, under the sponsorship of the National Research Council of Italy.

Many previous investigations have been undertaken with a view to determining appropriate test methods for evaluating the degree of suitability for welding of a generator, but hitherto results have not been decisive.

This paper summarizes the work so far accomplished in an extensive experimental investigation to evaluate the suitability for welding of a d.c. welding generator on the basis of particular dynamic behaviour which can be easily prescribed and, even more important, the results of which can be assessed without ambiguity.

The tests have been carried out using several d.c. arc welding generators having different characteristics. It is important to note that the majority of these generators are of types considered to be of medium or good performance in industrial service. In comparison with the work of most previous investigators, this study is exceptional both by the number of generators examined and by the selection of parameters which, as regards magnitude, correspond with present operating conditions of industrial welding generators.

It is essential to stress these points for a correct interpretation of the conclusions reached.

It should be noted that the static open-circuit voltage and the static values of short-circuit current do not appear to have any marked influence on the suitability for welding, but this merely signifies that, with modern welding generators, these parameters are within the limits beyond which their influence would be significant.

Correspondingly, this proposal for the criteria for s uitability for welding, considered alone and taken as the only basis of judgement, is valid only for welding machines at present available, because it is evident that, for an experimental welding generator which may have characteristics outside the above mentioned limits, these criteria by themselves cannot be considered sufficient.

However, because of the quality of machines which are available on the market, it does not seem worth while to complicate the judgement by the employment of supplementary criteria; that is why those criteria have not so far been introduced.

Finally, the fact that this work could be carried out on many d.c. welding generators having different welding characteristics made possible the determination of criteria which it would have been very difficult to establish in other circumstances.

Subjective Evaluation

In the first group of tests eight different generators were used. Practical tests were made by welding in the upwards position on plates 10 mm thick, arranged to form a butt joint, having a chamfer preparation in the form of a V with an included angle of 60°. This welding position was chosen because it provides clearly defined welding conditions and because it is considered to be the most difficult for maintaining an arc.

The welds were made from basic type electrodes 4 mm dia. specially manufactured to increase the difficulties of maintaining a stable arc.

The welding tests were made by several highly skilled welders having considerable practical experience. Each welder compared the performance of each generator with another until all possible combinations had been exhausted, and gave each time his judgement on the relative suitability for welding. It was then possible to establish for each welder a classification of generators on the basis of the subjective appreciation of their suitability for welding.

The classifications so established by the different welders varied somewhat, especially for machines of medium quality, but the best machines and the worst machines always retained their order of merit. On the basis of these different classifications an average classification was established which was retained as a basis of reference.

Influence of the Static Characteristic

No significant correlation could be established between this classification and the external static* characteristics of the generators.

Influence of the Performance of the Generators under Dynamic Conditions

The study concerning the dynamic behaviour of the generators began in the same way as those carried out by other investigators, *i.e.*, by the examination of the relevant oscillograms giving the voltage of the arc and the intensity of the current as a function of time during the course of welding operations. As is known, the most usual phenomenon to be examined is that which accompanies the transfer of a globule of molten metal from the electrode to the weld pool. During the transfer of the globule, the electrode and the workpiece are short-circuited by a piece of liquid metal and the voltage of the arc falls to zero.

This phenomenon is similar to that which occurs if the generator is feeding a load such as to give a voltage equal to the arc voltage, and if this load is then short-circuited for a period corresponding to the time of transfer of a globule of molten metal.

This is why it was at first thought that the tests could be effected with the aid of a switch, capable of being operated at a suitable speed and permitting the passage from a certain load condition either to total or partial short-circuit, and vice-versa, to provide data in relation to the suitability for welding of the generator.

Furthermore, for the first time (as far as is known by the author), arc power changes as a function of time were recorded by oscillograph during these tests.

The examination of numerous oscillograms made during tests carried out with a switch under the conditions described did not reveal characteristics leading to concrete results.†

On the contrary, the study of the oscillograms made during the welding operation showed the possibility of establishing a certain correlation between suitability for welding of the generators and the variations of arc power during the transient phenomena corresponding,

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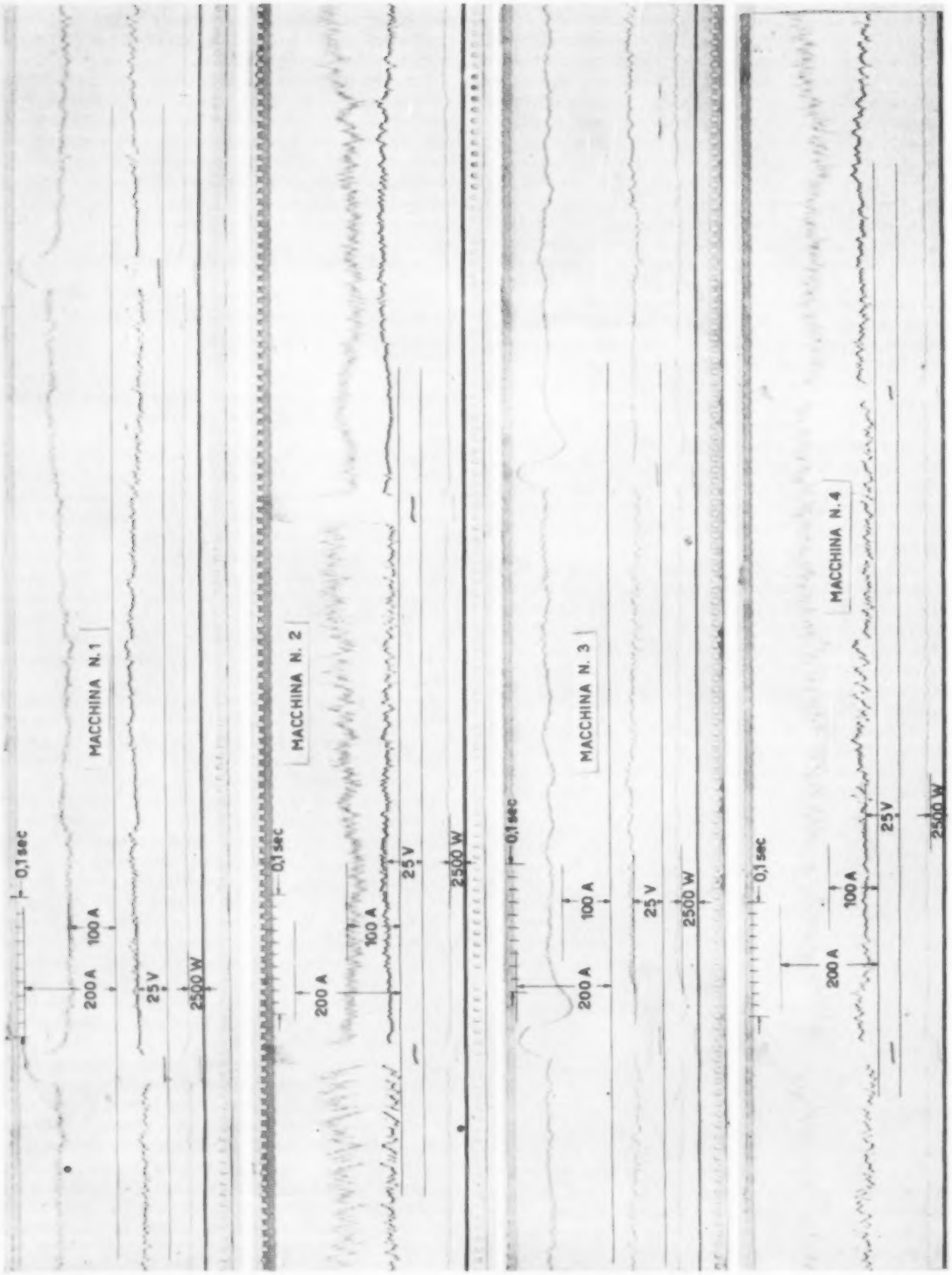
* Generator voltage under load expressed as a function of the load current.

† The dynamic characteristics as defined by Langkau and Jennings (see for example *Dynamic characteristics of d.c. welding machines*, 1950 Adams Lecture, *Welding J.*, February 1951) do not show any relationship with the classification established by the welders.


MACCHINA N. 1
0,1 sec
200 A
100 A
25 V
2500 W
MACCHINA N. 2
0,1 sec
200 A
100 A
25 V
2500 W
MACCHINA N. 3
0,1 sec
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MACCHINA N. 4
0,1 sec
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100 A
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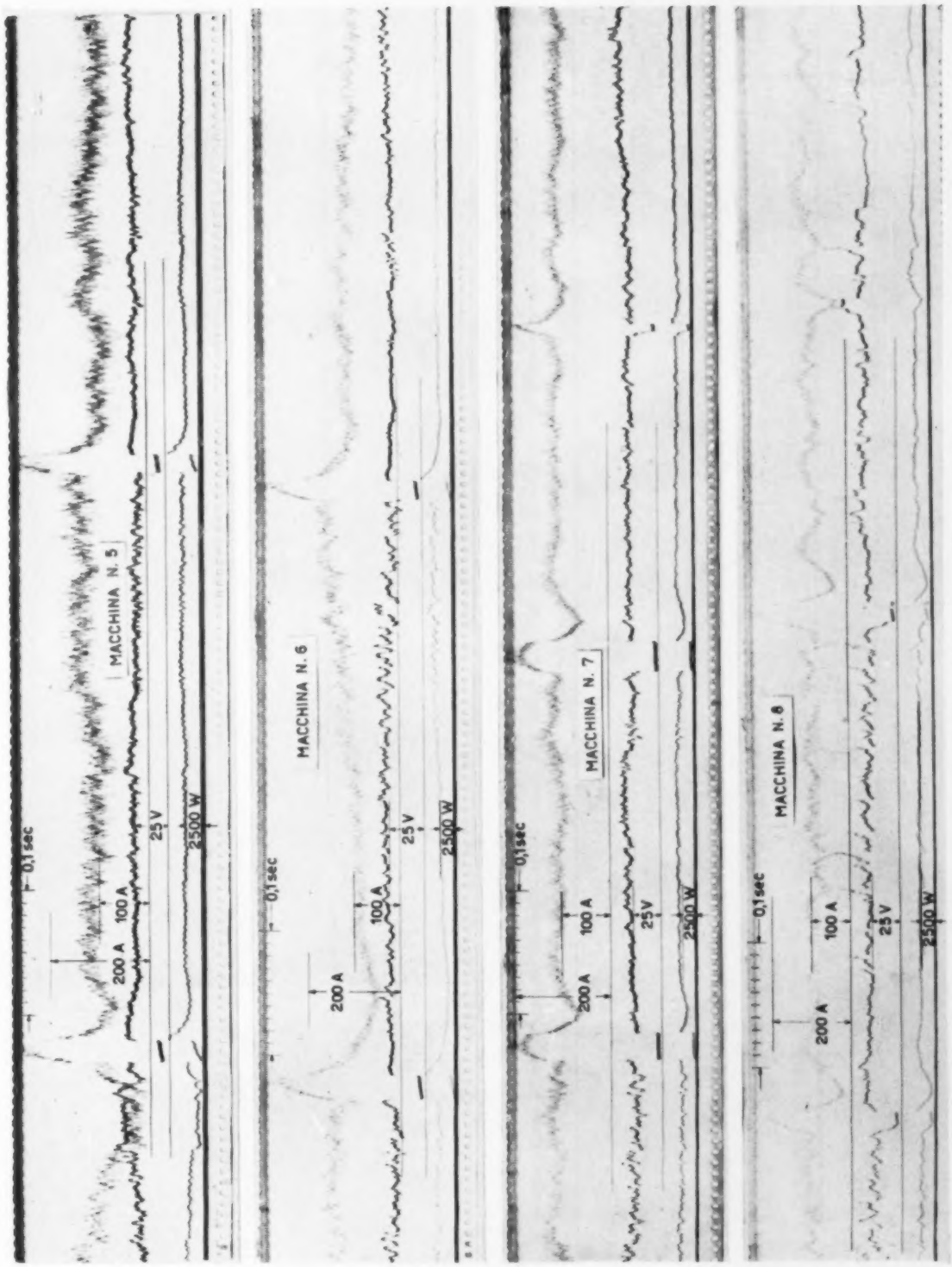
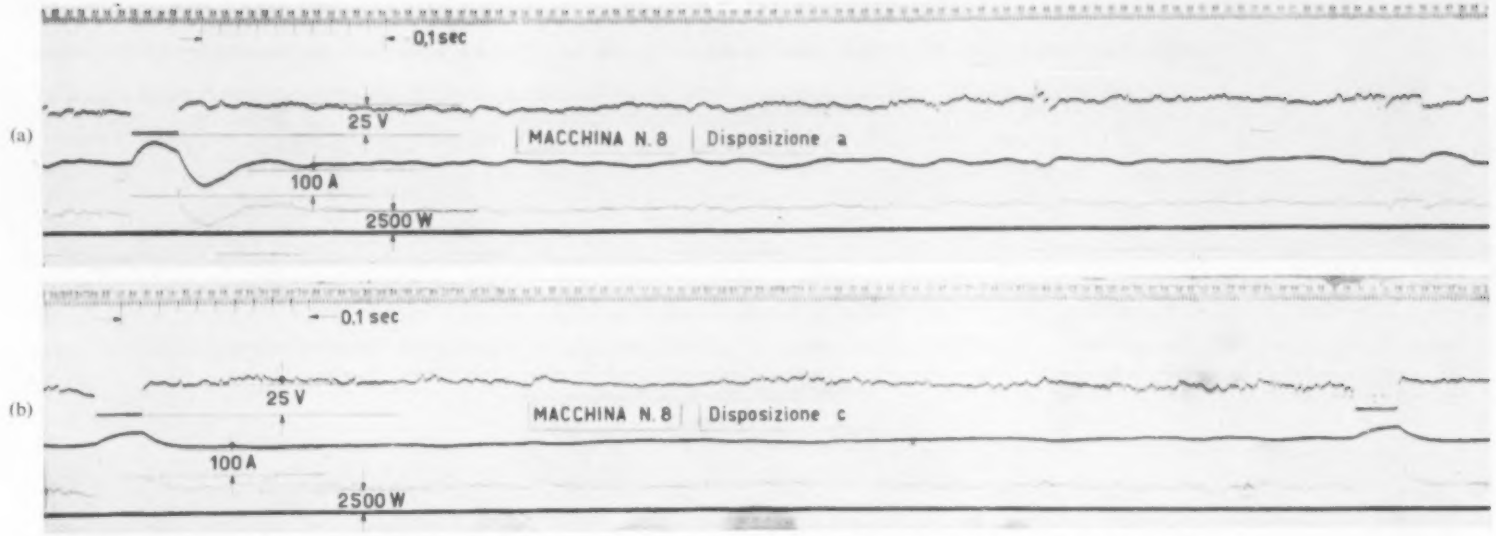
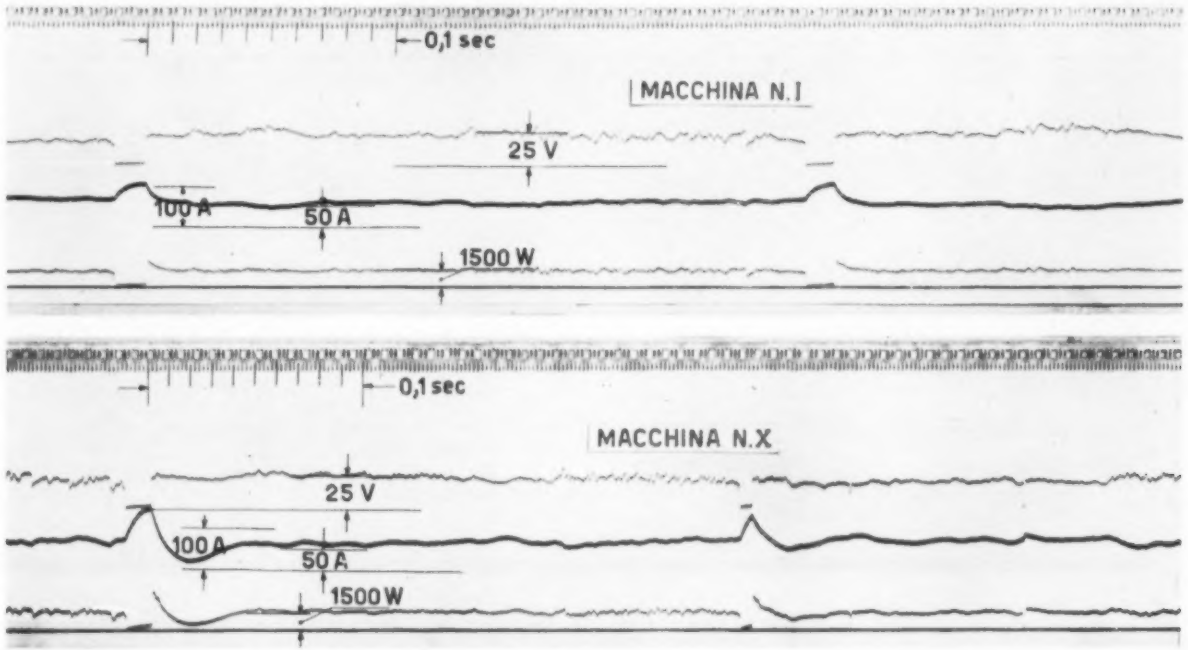


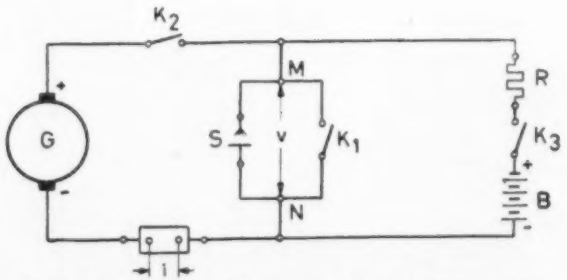
1—*Oscillograms of voltage, current, and power during welding, for the eight d.c. welding generators tested:* (a)–(h) *machines Nos. 1–8 as indicated*



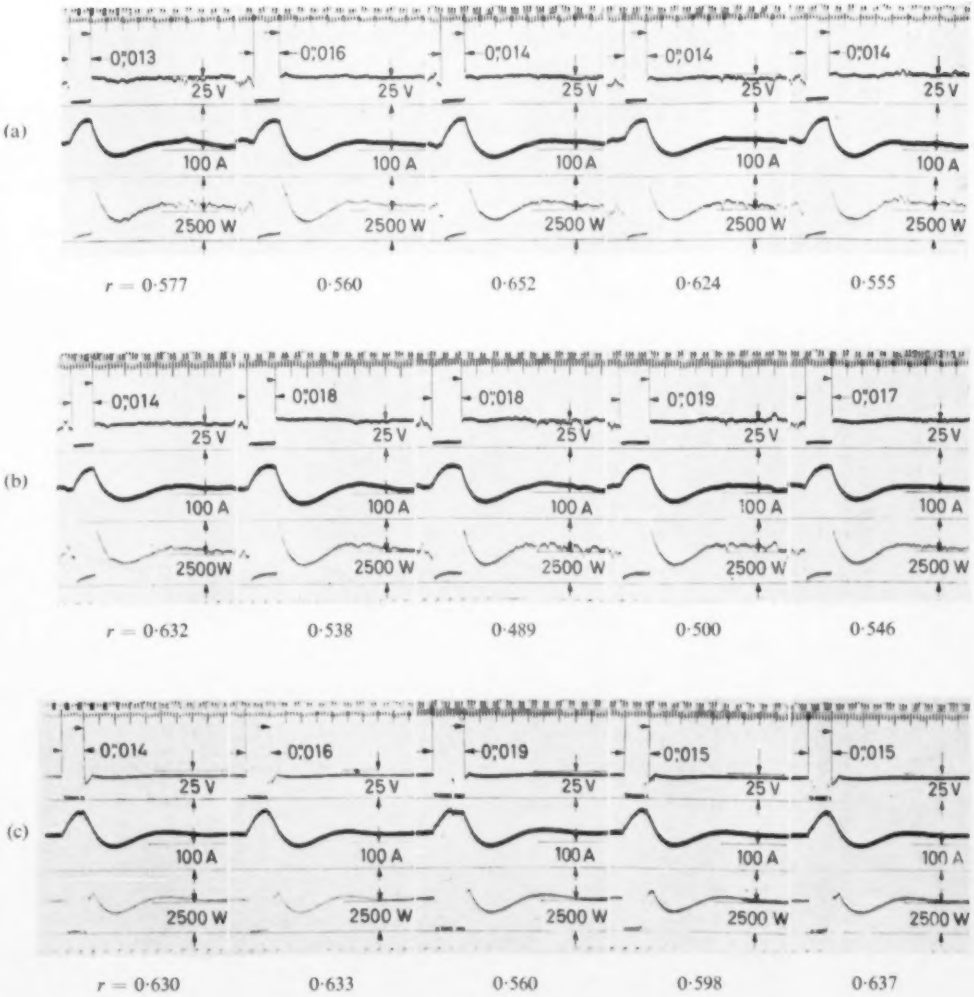
2—*Oscillograms of voltage, current, and power during welding with generator No. 8:* (a) *in its natural state,* (b) *improved with an inductance*



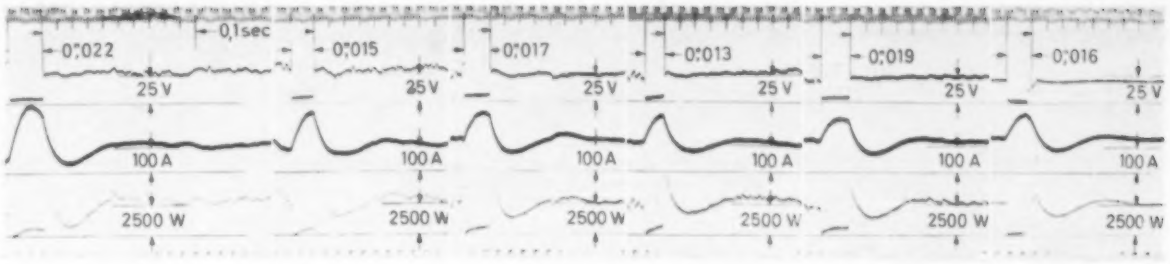
3—*Oscillograms of voltage, current, and power during welding with:* (a) *Machine No. I (ex No. 1).* (b) *Machine No. X (ex No. 8), using a low current (65–70 amp)*



4—*Electrical circuit used for comparative tests between short-circuits of drops and short-circuits obtained by a battery of accumulators and a switch*

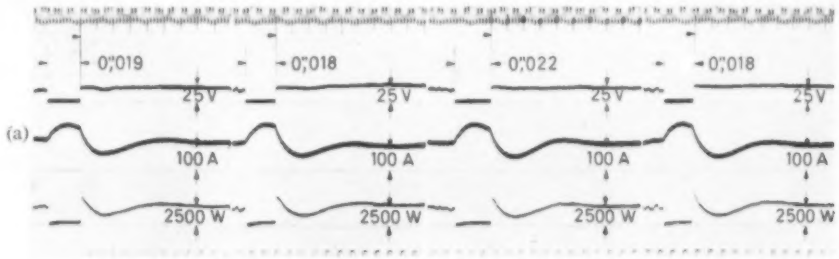


5—*Oscillograms of voltage, current, and power obtained for machine No. 8:* (a) *and* (b) *By normal welding operations,* (c) *by short-circuiting the generator with switch* K_1

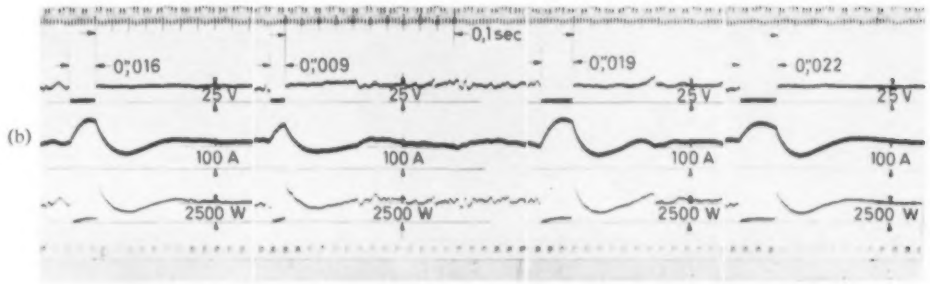


r = 0·274 0·437 0·499 0·596 0·517 0·625

6—*Oscillograms of voltage, current, and power for machine No. 8, showing drops obtained with electrodes of different types. (The last short-circuit, r=0·625, was obtained using the K_1 switch)*



r = 0·484 0·512 0·483 0·540



r = 0·537 0·682 0·547 0·447

7—*Oscillograms of voltage, current, and power for machine No. 8, showing drops obtained by welding*: (a) *Vertically upwards;* (b) *in the flat position*

as previously mentioned, to the transfer of a globule of molten metal from the electrode to the workpiece.

It was finally apparent, as the report shows, that, as the ratio between the minimum value of the power of the arc w, which occurs immediately after the passage of a globule of molten metal, and the average power of the arc under steady conditions W (*i.e.*, during the period between two drops of transfer when there are no short circuits between electrode and welding pool), decreases, so the suitability for welding of the generator declines.

The result seems to be logical because a lower value of the ratio $r = w/W$ corresponds to a lower energy provided to the arc on account of the transfer of the globule of molten metal. This lower availability of energy might correspond to the greater difficulty in maintaining an arc. In other words, for a given set of conditions, a deterioration in arc stability corresponds in some degree to a deterioration of the suitability for welding of the generator.

A first experimental confirmation of this assessment is obtained by examining the 8 oscillograms shown in Fig. 1. All these relate to welding operations carried out with basic electrodes of 4 mm dia. and a current of 115–120 amp, using the 8 machines previously submitted to subjective judgement. In the present study, the machines have been allotted reference numbers, in the order of the subjective classification, so that the machine judged to be best is machine No. 1 and that with the worst suitability for welding is machine No. 8. For these same machines, a quantitative evaluation of the value of r provides the following values:

Machine No.	*Ratio r*
1	0·975
2	0·970
3	0·685
4	0·940
5	0·735
6	0·735
7	0·620
8	0·463

A second experimental confirmation of this assessment was obtained from the oscillograms shown in Figs. 2*a* and 2*b*. These were taken in the course of welding operations with generator No. 8 in the natural state (Fig. 2*a*) and on the other hand with generator No. 8 improved as regards dynamic behaviour by the insertion of an inductance in the welding circuit (Fig. 2*b*).

In these two cases and under the same conditions of welding as previously mentioned the following values of r were determined:

Generator No. 8 (alone)	r = 0·463
Generator No. 8 (with an inductance in series)	r = 1·000

Subsequently, practical manual welding tests by different welders confirmed this improvement from the subjective point of view, generator No. 8 used with an inductance being judged equal or superior to generator No. 1.

A third experimental confirmation was obtained by testing ten generators or various welding circuits which included the majority of generators already mentioned. The method followed was the same in principle, but lower currents of the order of 65–70 amp were used. Figures 3*a* and *b* show the most typical oscillograms among those obtained.

Figure 3*a* is for machine No. I (ex No. 1), which was always classified as best by the subjective assessment by the welders. Figure 3*b* is for machine No. X (ex No. 8) which was always judged to be the worst.

For these generators, the following values were obtained for r:

Machine No. I	r = 1
Machine No. X	r = 0·362

Arising out of the first results obtained from this investigation, it has been suggested to Sub-Commission *B* that it should retain for the objective appreciation of the suitability for welding of a d.c. generator, the value of ratio r determined according to test conditions still to be defined in detail.

Further Studies and Confirmation

The results obtained in the course of these investigations gave rise to interesting discussions when presented to Sub-Commission *B*, under the Chairmanship of the French Delegate, Mr. Gaubert, at a meeting in Paris on 18th May 1957, and at the plenary session of Commission II at Essen in July 1957.

These discussions have given evidence of the opportunity to proceed with further work directed towards the adoption of the coefficient r as a means of defining the suitability for welding of generators.

Encouraged by the attention which the proposal had received, a new investigation was carried out with the object of confirming the proposal by as much experimental proof as possible. To obtain comparable results, the new tests were again carried out with 4 mm electrodes and at a current of 100–120 amp.

It was possible to establish the following points:

(*a*) The coefficient r may be determined by a method of test which does not require the services of a welding operator
(*b*) When the test is carried out as a welding operation the values obtained may be considered to have objective significance.

Point (*a*) was verified by comparative tests conducted on all machines and the values of r were found to correspond closely, as determined (i) by a welding operation and (ii) by putting the generators on short circuit in the following manner:

Adopting the circuit shown in Fig. 4 it is possible to:

(i) Weld, the arc being supplied with power from the generator G
(ii) Feed the battery of accumulators *B* from the generator *G* through the resistance *R*. For this purpose switches *K*2 and *K*3 are closed and switch *K*1, which has an automatic maximum current release, is open
(iii) Apply a short circuit for a short period of time between the points *M* and *N* by the application of automatic release switch *K*1, while the generator *G* is connected to the battery *B*.

This arrangement permits the study of the variations of voltage V between points *M* and *N* and the current i (measured through a shunt) which the generator is feeding to the battery of accumulators, when, for a short period of time, the generator is short-circuited, as is also the circuit comprising battery *B* and resistance *R*.

This latter resistance has the function of limiting the discharge current from the battery of accumulators when the short-circuit is applied between the points *M* and *N*.

Outside the periods of short-circuit the generator charges the *B* battery, its back e.m.f. representing the arc voltage. The duration of short-circuit obtained by the operation of *K*1 was exactly comparable to that of a short-circuit in a welding operation caused by the passage of a globule of molten metal from the electrode to the welding pool.

Figures 5*a* and *b*, which relate to the same generator, make possible an appreciation of the bases on which a comparison could be established.

Furthermore, the comment under point (*b*) is justified by the fact that, for each machine, the oscillograms taken in the course of fusion of various types of basic electrodes have led to similar values of the coefficient *r*. Figure 6 gives an example of the comparison obtained from one of the tested generators

These oscillograms are actually a very small part only of those which were taken. In fact, they were made following other studies which had proved that the phenomena concerning the transference of a globule of molten metal from the electrode to the welding pool are independent of the welding position.

These studies were made with special reference to oscillograms taken during the course of welding operations in two fundamental positions and in conditions of deposition which are very typical:

(*a*) Making a first run in a single V joint in the vertical upwards position where the maximum difficulty is experienced

(*b*) Making a simple deposit in the downhand position, which is the easiest welding operation.

The relevant oscillograms obtained in these two welding positions are analogous as is shown by the examples given in Fig. 7.

Finally, as an indication from the point of view of statistics, 470 oscillograms were quantitatively examined during this investigation.

Consequently, it can be claimed that the proposal to define the suitability for welding of a d.c. arc welding generator in terms of *r* within the meaning previously defined, is based upon a mass of experimental facts which are significant.

Another conclusion, which is associated with the main one and does not detract from it, may also be deduced from the observations made: this secondary conclusion enables a simpler test method to be used which in turn makes possible another valid definition of suitability for welding.

During the examination of the results of the first group of tests by Sub-Commission B, it was suggested that the coefficient ρ could possibly be substituted for the coefficient *r*, where ρ is defined as $\rho = i/I$. In this expression i is the minimum value of current at the instant following the passage of a globule of molten metal and I is the average value of arc current under steady conditions.

An examination of the relevant oscillograms shows evidence of the possibility of reverting to the coefficient ρ, but it should be noted that, at low values of *r*, the corresponding values of ρ are slightly higher. Judging by the present experiments, the difference between the value of *r* and of ρ is approximately 0·1 for generators having a particularly low value of the coefficient *r*.

However, it is clear that when the value of ρ obtained makes the appreciation of suitability for welding uncertain, reference could be made to the corresponding value of the coefficient *r*, which is in direct correlation with the fluctuation of power of the arc, and which is better adapted for the purpose in view.

As all the experiments have been made with low current intensities, a comment could be made in this connection. However, the choice of the range of current examined is considered to be justified because the stability of the arc, independently of the fluctuations of power, is always better when the values of the welding current are higher.

Conclusions

As a conclusion to this study, it is proposed to define the suitability for welding of a d.c. arc welding generator in terms of the coefficient *r* and, less precisely, by the coefficient ρ, the values of these coefficients being taken as the average resulting from 10 determinations made on an oscillograph and obtained during deposition of a basic electrode on a flat steel plate.

These 10 determinations relate to the transfer of 10 globules of molten metal, each giving rise to a definite short-circuit. The average values thus obtained are practically independent of the skill of the operator and of the type of basic electrode used. It is established that these factors are significant.

The suitability for welding would be considered sufficiently satisfactory if:

$$r > 0{\cdot}6 \text{ or if } \rho > 0{\cdot}7$$

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Welding

RESEARCH

B.W.R.A. REPORT

The Effect of Oxygen Additions to the CO_2– Shielded Metal Arc

By A. A. Smith, B.SC., A.I.M.

The effect of adding excess oxygen to the CO_2 shielding gas in the inert-gas metal-arc process has been investigated. The burn-off rate and droplet transfer frequency of the process are not affected by the addition of up to 10% oxygen. In larger amounts, up to 100%, the droplet transfer rate is still unaffected but gross porosity occurs, depending on the deoxidizing qualities of the filler wire.

THE addition of 1–5% flow of oxygen to argon is now standard practice for the inert-gas metal-arc welding of steel. It improves the operating characteristics of the process by reducing the size of the droplets of material transferred across the arc and increasing their frequency of transfer. The transfer of material as small molten globules from the electrode wire tip is believed to be largely due to the magnetic pinch effect, and the effect of adding oxygen to the shielding gas would be to reduce the surface tension at the neck of the transferring droplet and assist in its detachment. When CO_2 is used as the shielding gas, it has been found that the metal is transferred across the arc in the form of large globules and the welding characteristics are not ideal. Although oxygen would be expected to be present in the arc atmosphere as a result of dissociation of CO_2, the evidence suggests that the ease of detachment of droplets in this oxidizing gas is inferior to that in argon containing only minor amounts of oxygen. The present work has investigated the effect of adding an excess of oxygen to the CO_2 shielding gas by measuring the droplet transfer rate of a standard welding wire over a wide range of welding current.

Experimental Technique

A standard testing technique was adopted using controlled-arc welding equipment and a 600 amp transformer-rectifier power source. Weld beads were made on $\frac{1}{2}$ in. thick mild steel plate using a mechanised traversing unit on which constant welding speeds could be preset. Current, voltage, and weld time were recorded automatically, and wire feed speed was computed from wire consumption read from a revolution counter mounted on the wire drive rolls. A double-beam oscilloscope was used to record detail in the arc voltage wave form over short periods. The oscillograms were analysed to provide the metal transfer frequency data, individual transfers being indicated by a typical 'saw-tooth' wave form or voltage peaks.[1]

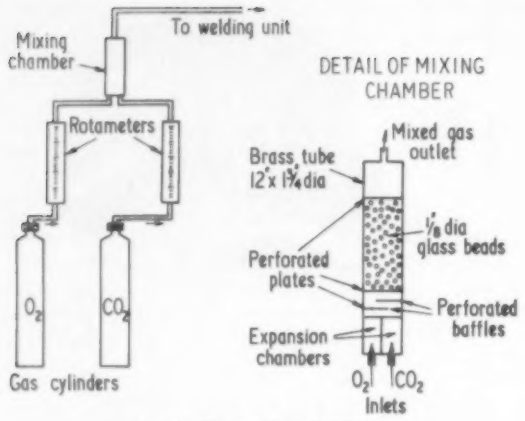
Table I

Composition of filler wires

	C, %	Si, %	Mn, %	Al, %	Zr, %	Ti, %
MS65 (British)	0·05	0·62	1·4	0·12	0·6	0·15
SV5 (German)	0·10	1·2	1·4	0·01	n.a.	trace

The test beads were made using $\frac{1}{16}$ in. dia. MS65 wire (Table I) at an arc voltage of 30 V to avoid audible short circuiting and thus make oscillograms easier to assess for metal transfer data. Care was

Report A1/1/58 of the British Welding Research Association circulated to members in April 1959.
Mr. Smith is a Principal Scientific Officer with the Association.



1—*Apparatus for mixing gases*

taken to ensure the proper mixing of the two gases by the use of the baffle and mixing chamber shown in Fig. 1. Accurate metering of the gases was possible by means of the calibrated rotameters.

The test constants were: Ratio of current (amp) to travel speed (in./min) 10 to 1, giving an approximate energy input to the plate of 17·5 kW sec/in.; gas flow 35 cu.ft/hr; torch angle 90°; ¾ in. dia. nozzle, ½ in. from plate.

Transverse macrosections, measured for melted plate area at a magnification of 10 were taken at central portions of the weld beads coincident with the oscillographic recordings.

Experimental Results

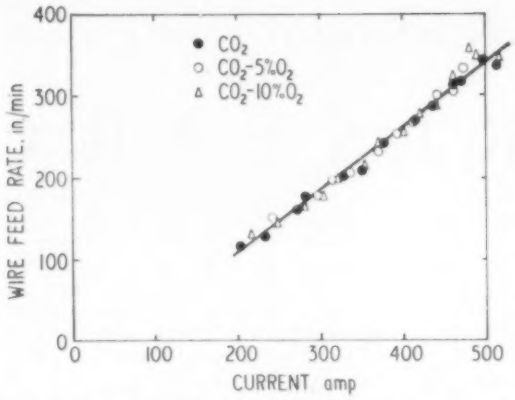
Burn-off characteristics of $\frac{1}{16}$ in. dia. MS65 wire in CO_2 and CO_2-O_2 mixtures

The relationship between current and burn-off rate at the test voltage of 30 V is plotted in Fig. 2 for tests made in CO_2 and with the addition of 5% and 10% oxygen. There is no significant change of burn-off rate by the addition of oxygen, and all three test series may be represented by the same curve.

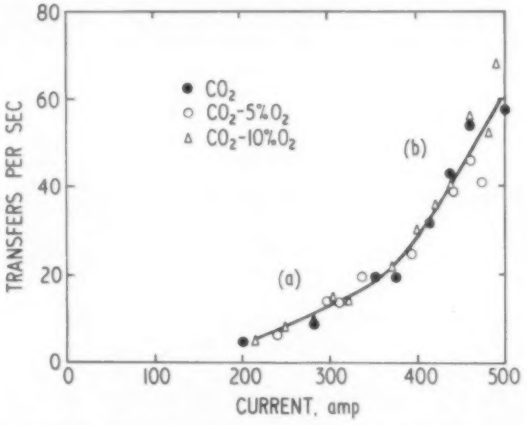
Oscillographic data

The simple oscillographic technique used in these tests enabled the incidence of metal transfer to be counted from the pronounced voltage flicks or discontinuities which occur on the voltage wave form. The 300 cycle 3-phase power source ripple was also evident but did not prevent the interpretation of the records. The effect of current on the metal transfer rate is shown in Fig. 3 for the pure CO_2 and CO_2–O_2 mixtures. The transfer rate of droplets increases with current, and experimental points for tests made with shielding gas containing 5 or 10% oxygen are not appreciably different from those made with pure CO_2.

Three types of oscillograms were obtained. The first consisted of a saw-tooth wave form with pronounced voltage peaks at regular intervals. These records were obtained only for tests made at the lower currents (200–380 amp) and were similar to the oscillograms obtained in the sub-threshold range at low current in the aluminium arc in argon.[1] An example of this type



2—*Burn-off characteristic of $\frac{1}{16}$ in. dia. MS.65 wire in CO_2, CO_2–5% O_2 and CO_2–10% O_2. 30 V arc*



3—*Droplet transfer rate for tests in CO_2, CO_2–5% O_2, and CO_2–10% O_2. $\frac{1}{16}$ in. dia. MS.65 wire, 30 V arc*

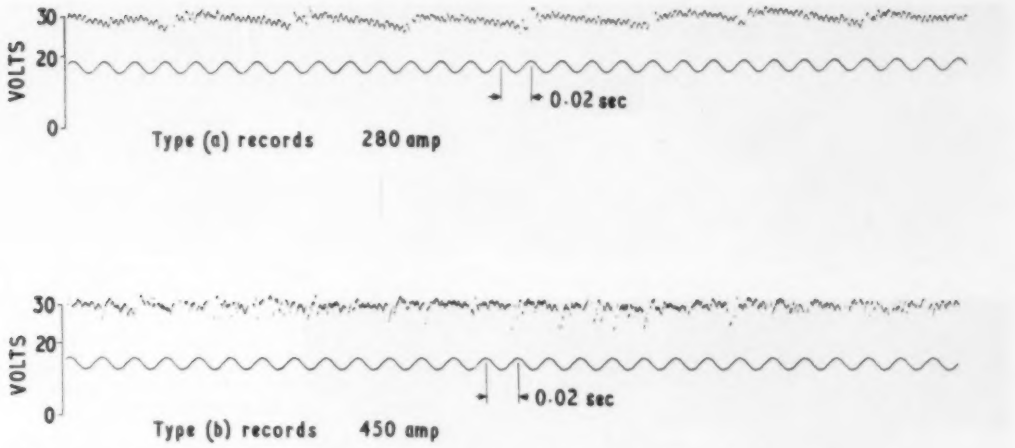
of record, which was confined to the part labelled (*a*) in Fig. 3, is shown in Fig. 4 (type *a* record).

The second type of oscillogram exhibited a more rapid fluctuation without the saw-tooth waveform, and was typical of tests made at currents of 400–500 amp. These records were confined to the steeper part of the curve (*b*) in Fig. 3. An example of this type of record is shown in Fig. 4 (type *b*).

The oscillograms for beads made at currents above 500 amp consisted of irregular complex oscillations with flicks towards zero voltage, and no material transfer measurements were possible.

For a given test current, however, there was no difference in the type of oscillogram or wave form, and the examples shown in Fig. 4 may be regarded as typical for all three gas mixtures.

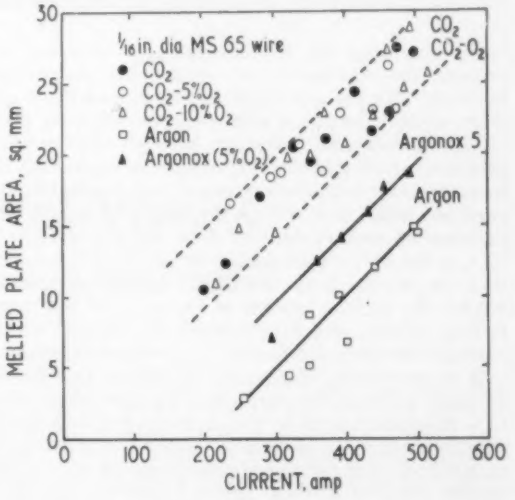
The oscillographic technique did not enable a quantitative comparison to be made of the metal transfer frequencies in argon and argon-oxygen mixtures, because the 3-phase ripple due to the power source made the records difficult to interpret. At a similar current however, the argon-5% oxygen shielding gas produced a much finer ripple of sharp voltage peaks than either plain argon or the CO_2 mixtures.



4—*Oscillograms of arc voltage showing metal transfer typical of pure CO_2 or CO_2–O_2 mixtures*

Examination of bead sections

Transverse sections taken from the test beads were traced at a magnification of 10, and the results are shown graphically in Fig. 5 together with comparative results for argon and argon-5% oxygen. The scatter in results for the tests in CO_2 is apparent, but the addition of oxygen does not appear to alter the melting efficiency of the arc, and all the results are contained in the dotted area. Typical macrosections are shown in Fig. 6, where the effect of changing the shielding gas at constant current and welding speed is shown. The three macrosections (*a*), (*b*) and (*c*) are the bead shapes for tests made in CO_2 containing 0, 5, and 10% oxygen respectively. There is virtually no difference in either the profile or the amount of the melted area. The comparative appearance of the beads made in argon and argon–5% oxygen is shown in the macrosections (*d*) and (*e*) respectively, and the small area of melted plate contrasts markedly with the tests in CO_2. The characteristic 'finger' penetration associated with argon–oxygen shielding shown in macrosection (*e*) is always absent in the beads made in CO_2.



5—*Effect of current on area of melted plate in various shielding gases*

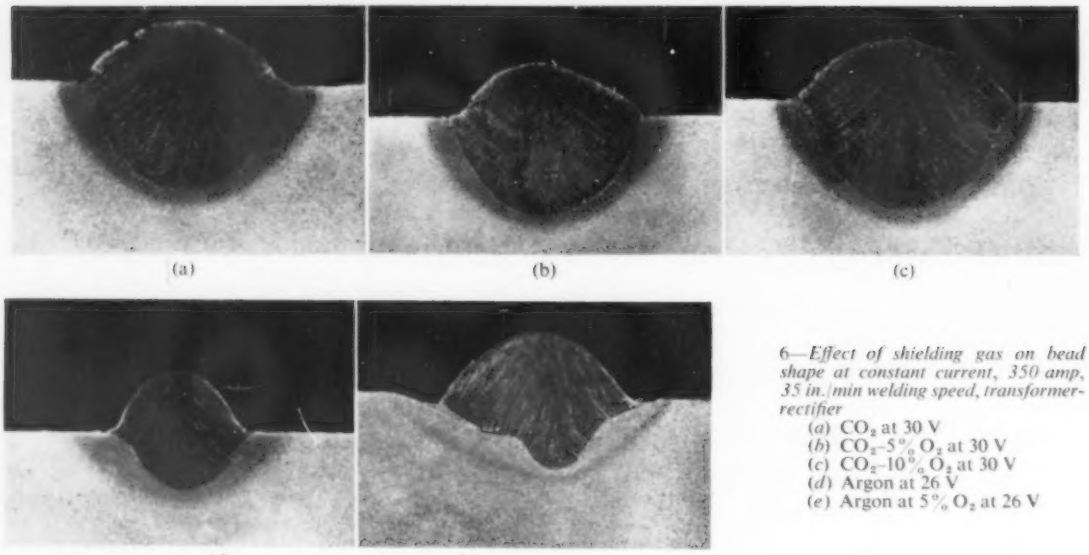
Discussion

Addition of up to 10% oxygen to CO_2

The results of this investigation show that the addition of up to 10% oxygen to the CO_2 shielding gas neither affects the wire burn-off rate nor the frequency or size of the metal transfer. The addition of oxygen has been shown to increase the ductility[2] and reduce the hydrogen content[3] of weld metal in CO_2-shielded welding, but no alteration in the metal transfer characteristics may be attributed to this change in the composition of the shielding gas.

It is surprising that the coarse 'sub-threshold' globular type of transfer can exist up to nearly 400 amp, irrespective of oxygen addition. The metal transfer frequency ranges from 5 transfers per sec at 200 amp to about 25 transfers at 380 amp; from a knowledge of the wire feed rate it was calculated that these globules contain 0·4 in. of $\frac{1}{16}$ in. dia. wire at 200 amp and about $\frac{1}{16}$ in. of wire at 380 amp. Globules of metal collected from the 200 amp tests were weighed and confirmed the oscillographic estimation of size. With the aluminium arc in argon, $\frac{1}{16}$ in. of wire per drop is on the threshold for 'spray-type' transfer with $\frac{1}{16}$ in. dia. wire, but occurs at 130 amp. The transfer frequencies for the aluminium arc in argon at 200 and 380 amp would be 250 transfers per sec and 500 transfers per sec, respectively, when fine 'spray' type transfer without spatter would be obtained.

In the CO_2-shielded arc, where a coarse mode of material transfer occurs up to high currents, it is not surprising that excessive spatter can occur, especially since the globules of metal are not transferred consistently in line with the electrode. Spatter occurred with all the test beads made below 380 amp exhibiting type (*a*) oscillograms.



(a) (b) (c)

(d) (e)

6—Effect of shielding gas on bead shape at constant current, 350 amp, 35 in./min welding speed, transformer-rectifier
(*a*) CO_2 at 30 V
(*b*) CO_2–5% O_2 at 30 V
(*c*) CO_2–10% O_2 at 30 V
(*d*) Argon at 26 V
(*e*) Argon at 5% O_2 at 26 V

By operating the low-current arcs at lower arc voltage the occurrence of spatter can be limited because, with the arc below the surface of the work, there is more chance of entrapment in the weld pool.

The fundamental difference between the beneficial effects of small additions of oxygen in improving the metal transfer behaviour in argon and the inability to exert any influence in CO_2 is not fully understood. The presence of oxygen derived from the dissociation of CO_2 in the arc, supplemented by up to 10% by addition as in the tests described, cannot apparently reduce the surface tension at the neck of the transferring droplet and facilitate its detachment in the manner visualized for the effect of oxygen in argon.

It is reasonable to assume, therefore, that fundamentally different arc characteristics exist when either CO_2 or argon is used as the shielding gas, and special experimental work is clearly required to provide an understanding of the actual mechanism of transferring molten droplets in these gases. To obtain this data, refined techniques involving high speed cine photography, spectrography, and surface tension measurements will probably be necessary.

Major additions of oxygen (up to 100%) to CO_2

To determine whether the presence of amounts of oxygen greater than 10% in the CO_2 shielding gas would have an effect on the droplet transfer rate, a further series of tests was carried out. Bead-on-plate weld runs were made using $\frac{3}{32}$ in. dia. MS65 wire and, the shielding gas contained different amounts of oxygen from 0% to 100%. The constant welding conditions used were: Current 520 amp, voltage 30 V, welding speed 26 in./min. The wire burn-off rate and droplet transfer rate were determined as before and the results are given in Table II.

The addition of as much as 75% oxygen resulted in only a slight increase in the burn-off rate over that in plain CO_2 (107 to 119 in./min), but in 100% oxygen there was a marked increase to 144 in./min, representing a 35% increase over CO_2.

The droplet transfer rate was almost constant at the test current of 520 amp irrespective of the wide variation of oxygen content in the shielding gas composition. At this current, the reduction in the droplet transfer rate, using $\frac{3}{32}$ in. wire, to 14 transfers per sec compared with 60 transfers per sec with $\frac{1}{16}$ in. wire in CO_2, should also be noticed.

Table II

Effect of oxygen additions to CO_2 shielding gas on wire burn-off rate and droplet transfer rate

Oxygen in CO_2, %	0	12	20	25	30	41	54	71	75	100
Wire feed rate, in./min	107	104	109	106	110	109	115	117	119	144
Transfer frequency, transfers per sec	14	13	12	14	11	14	13	14	14	13·5

The weld beads made with more than 50% oxygen exhibited gross open surface porosity, whilst internal porosity was revealed in beads made with more than 25% oxygen.

Tests with a German welding wire (Table I) which had the same manganese content as the British wire but with 1·2% Si compared with 0·6% Si (and Al, Zr, Ti) showed surface porosity only above 80% oxygen addition and internal porosity above 50% oxygen.

Conclusions

(1) The addition of up to 10% oxygen to the CO_2 shielding gas does not alter the burn-off rate of $\frac{1}{16}$ in. MS65 wire. The rate of transfer of metal droplets across the arc is also unaffected by the addition of oxygen to the CO_2 shielding gas.

The existence of the coarse 'sub-threshold' type transfer, as evidenced by the typical saw-tooth oscillogram of arc voltage, has been observed for currents up to nearly 400 amp. The use of higher

currents results in a smaller droplet size, accompanied by a reduction of spatter. No estimate of transfer rate is possible for beads made above 500 amp with the present technique because of the complicated arc-voltage waveform, but the records suggest frequent metal transfer.

Comparative tests made on the same welding wire in argon and argon–5% oxygen, indicate a more regular and rapid transfer behaviour in the argon–oxygen mixture. For the same current, the area of welded plate is greater with CO_2 shielding than with argon–oxygen.

(2) The addition of up to 75% oxygen to the CO_2 shielding gas, using $\frac{3}{32}$ in. dia. MS65 wire, results in only a slight increase in burn-off rate. In pure oxygen the burn-off rate is increased some 35% over that in pure CO_2. No difference is observed in the metal transfer frequency at a given current (520 amp) with this wire over a range of oxygen additions from 0 to 100%.

The present work shows therefore that the basic operating characteristics of the CO_2-shielded metal arc (burn-off rate and material transfer frequency) are unaffected by the addition of up to 10% oxygen. The addition of larger percentages of oxygen to the CO_2 shielding gas (up to 100%) has no effect on the droplet transfer rate, and results in gross porosity above a certain level, depending on the amount of deoxidant in the filler wire.

Acknowledgments

The author wishes to acknowledge the assistance of Mr. A. G. Brain, B.A., A.I.M., and Mr. G. A. Phipps, A.Inst.W. with the experimental work.

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B.W.R.A. REPORT

Stresses in Cylindrical Vessels

DUE TO LOCAL HEATING STRESS RELIEF OF CIRCUMFERENTIAL WELDS

By R. T. Rose, B.SC., PH.D. A.M.I.MECH.E.

In heating a pipe or vessel locally to a circumferential butt weld, with the object of stress-relieving the weld, there is a danger that the local heating may itself result in the appearance of residual stresses. The present approximate analysis indicates the level of thermal stress to be expected, as a function of the length of pipe heated and the pipe mean radius and thickness.

WHERE it is not feasible to stress-relieve a weld by heating the entire structure, an attractive alternative is to apply heat only in the vicinity of the weld. This note attempts to provide a rational basis for the selection of the heating conditions for the stress-relief of a complete circumferential weld in a plain cylindrical shell.

Because localized heating is bound to induce thermal stresses, it is possible that a secondary residual stress system may be set up if the thermal stresses at any time reach the hot yield strength of the material. The need arises, therefore, to control the temperature distribution in such a way as to limit the thermal stresses to a safe level.

The following analysis of the thermal stresses due to an arbitrary temperature distribution (which is nevertheless believed to be representative) enables the stresses to be estimated for any given heated length of shell, and so permits the choice of the latter for satisfactory stress-relief. It is assumed that the heat is applied uniformly to a band around the vessel so that axial symmetry is maintained.

Temperature Distribution

A solution of the problem of the heat flow in an infinite plate due to the input of heat to a band of specified width and infinite length is given by Carslaw and Jaeger.[1] Since the vessel has axial symmetry, the problems are seen to be identical. The solution is in terms of the rate of heat input (constant), the width of the heated band, the distance of a point at which the temperature is required, and the time of heating. It is

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Dr. Rose is a Principal Scientific Officer with the Association.

assumed that there is no loss of heat from the plate surfaces.

Unfortunately the functions in which the transient temperature distribution is expressed have, so far, led only to a laborious solution for the thermal stresses. It has not been found possible to reduce the number of variables so as to make the work practicable for general application, and recourse has been made to an approximate stationary temperature distribution in terms of functions amenable to simple solution.

This approximate temperature distribution is

$$T = T_o \exp(-ax)(\cos ax + \sin ax)$$

where T is the temperature of a point distant x from the centre of the heated band at which the temperature is T_o, and a is a parameter depending on the rate of heating and the width of the heated band. The form of this function is such that the temperature gradient is zero at the centre of the heated band and again zero at large distances from the band, where the temperature itself is zero. Negative temperatures appear at smaller distances, but they are never greater than 5% of the maximum temperature, and are considered to be of small importance.

The validity of this approximation is limited, because adjustment of the single parameter a is not sufficient to cover the variety of temperature distributions obtainable by varying the heating conditions. However, the following reasoning shows that the approximation is satisfactory for practical cases. If it is assumed that the rate of heat input per unit area of heated band is constant, then the total rate of heat input increases with the width of the band. An analysis of Carslaw and Jaeger's solution shows that, as might be expected, the time required to heat the centre of the band to the stress-relieving temperature is very large for a narrow band, and decreases to a constant minimum value when the band width is increased. At the same time, the total energy consumed in the heating decreases to a minimum value (at a band width slightly less than that at which minimum time is reached) and thereafter increases roughly linearly with band width.

Making the assumption that it is preferable to employ the shortest heating time, even though this entails the consumption of slightly more energy than the minimum, the type of temperature distribution curve that may be expected is liable to only small variations. In particular, the temperature at the edge of the heated band will be a little over one-half of the temperature at the centre. If the parameter a in the approximate expression for temperature is chosen such that $ax_h/2 = 1{\cdot}0$, where x_h is the width of the heated band, the temperature at the edge of the band is $0{\cdot}51\ T_o$. Further, the second derivative of the temperature distribution curve passes through a zero value at $ax = 0{\cdot}8$; *i.e.*, very near the edge of the band. In fact, the exact temperature distribution shows that the second derivative of temperature with respect to distance is discontinuous at $x = x_h/2$ at which point it changes from a maximum negative value to a maximum positive value. The approximate expression is continuous and therefore the boundary at $x = x_h/2$ has been made to disappear, with considerable simplification to the stress solution. The justification for this simplification rests on two considerations: Firstly, that the approximate temperature distribution gives the temperature at a point with small error; secondly, that the boundary is sufficiently removed from the centre of the heated band $x = o$, where the greatest stresses may be expected, for discontinuity stresses arising from the change in second derivative of temperature with respect to distance to have little effect.

Stress Distribution

The method of calculating stresses arising from a given temperature distribution is as follows.[2]

Firstly, it is supposed that the shell is constrained at all points, so that thermal expansion is suppressed entirely in the hoop direction while longitudinal expansion is unrestrained. Hoop stresses of magnitude $E\alpha T$ (where α is the coefficient of linear expansion and E is Young's modulus) are introduced. An external radial pressure of value $E\alpha T\,(t/R)$ is necessary to produce this condition, where t is the thickness of the vessel and R its mean radius. This pressure is constant round a circumference and varies with the temperature in the axial direction. Since there is no deflection of the tube in this condition, no stress other than the hoop compression $E\alpha T$ is present.

Secondly, it is supposed that an internal pressure of value $E\alpha T\,(t/R)$ is acting alone on the tube at room temperature. This causes a radial deflection of the tube, varying in the axial direction and giving rise to bending and direct stresses. Superimposing the two systems produces the case in which the tube is unloaded but subjected to a non-uniform temperature distribution, with stresses corresponding to the algebraic sum of the two systems.

The radial deflection w of the tube in the second system is governed by the equation:

$$\frac{d^4w}{dx^4} + 4\beta^4 w = E\alpha T \cdot \frac{t}{R} \cdot \frac{1}{D}$$

where

x = axial co-ordinate

$\beta^4 = \dfrac{3(1-\nu^2)}{R^2 t^2}$

$D = \dfrac{Et^3}{12(1-\nu^2)}$ = flexural rigidity

ν = Poisson's ratio

This may be written:

$$\frac{d^4w}{dx^4} + 4\beta^4 w = 4\beta^4 \cdot R\alpha T$$

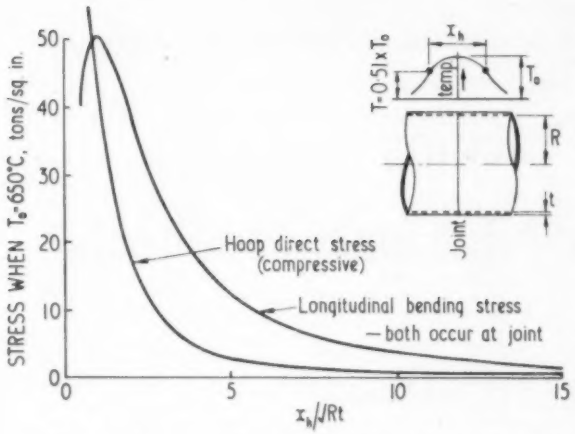
and when the approximate temperature function is introduced becomes:

$$\frac{d^4w}{dx^4} + 4\beta^4 w = 4\beta^4 \cdot R\alpha T_o \exp(-ax)(\cos ax + \sin ax)$$

The solution is

$$w = C_1 \exp(-\beta x)\cos \beta x + C_2 \exp(-\beta x)\sin \beta x + \frac{\beta^4}{\beta^4 - a^4} R\alpha T_o \exp(-ax)(\cos ax + \sin ax)$$

where C_1 and C_2 are arbitrary constants chosen to suit the boundary conditions at $x = o$, x being measured from the centre of the heated band. Two further terms in the solution have been omitted so as



1—*Relationship between heated length* x_b *and maximum stresses induced, assuming no yielding*

to satisfy the boundary condition as x approaches infinity and w tends to zero. At $x=0$ the boundary conditions are zero slope and zero shear force; that is, the first and third derivatives of w are both zero. From these conditions the arbitrary constants in the solution are calculated and the expression becomes:

$$w = R\alpha T_0 \frac{\exp(-ax)(\cos ax+\sin ax)-k^3\exp(-\beta x)(\cos\beta x+\sin\beta x)}{1-k^4}$$

where $k=a/\beta$

Making use of the relations hoop direct force $N_\theta=\frac{Et}{R}w$ and axial bending moment $M=-D.\frac{d^2w}{dx^2}$, both per unit length of section, and adding the hoop stress for the system having full restraint of thermal expansion, the final stresses become:

$$\sigma_\theta = E\alpha T_0 \frac{k^3}{1-k^4}\cdot\left\{k.\exp(-ax)\,(\cos ax+\sin ax)-\exp(-\beta x)\,(\cos\beta x+\sin\beta x)\right\}$$

$$\sigma_b = 1{\cdot}83\,E\alpha T_0 \frac{k^2}{1-k^4}\cdot\left\{\exp(-ax)\,(\cos ax-\sin ax)-k.\exp(-\beta x)\,(\cos\beta x-\sin\beta x)\right\}$$

where σ_θ is the direct hoop stress and σ_b is the axial bending stress, and Poisson's ratio has been taken as 0·3. The hoop direct stress is accompanied by a bending stress of value $0{\cdot}3\,\sigma_b$.

The largest values of these stresses occur at the point of maximum temperature and are therefore of most significance. Their values are:

$$\sigma_\theta = -E\alpha T_0 \frac{k^3}{(1+k)(1+k^2)}$$

$$\sigma_b = \frac{-1{\cdot}83}{k}\,\theta$$

at $x=0$.

Adopting the choice of the parameter a as previously proposed, it is found that:

$$k=\frac{a}{\beta}=\frac{2}{x_b}\cdot\frac{\sqrt{(Rt)}}{1{\cdot}286}=1{\cdot}56\,\frac{\sqrt{(Rt)}}{x_b}$$

The stress components when $T_0=650°C.$ are plotted in Fig. 1 with $x_b/\sqrt{(Rt)}$ as abscissa, making the assumption that the material remains elastic throughout, and using a representative value for the coefficient of linear expansion of low-carbon steel.

Residual Stresses

During the heating, the yield strength of the material diminishes while the thermal stress increases. Should the thermal stress exceed the value which the material can sustain, some local yielding will occur and a redistribution of stress will take place. Provided that the temperature distribution remains of similar form during cooling, so that each point of the material passes through the predicted thermal stress range in a straightforward manner, no further yielding of the material can occur (except in the extreme case of reverse yielding at or near room temperature). The residual stress may then be expected to be approximately equal to the difference between the (elastic) thermal stress range predicted at the point of maximum temperature and the yield strength of the material at this temperature.

Whilst it is necessary to limit the residual stresses, it is difficult to decide the level at which they become intolerable. Further, it is possible that the bending stresses are of less importance than the direct stresses. Because of these uncertainties and the approximate nature of the assumed temperature distribution, the results given here would serve their best purpose if they were used as a scaling relationship, to determine conditions of stress relief for a vessel of new proportions from those found satisfactory with other vessels in the past.

The accumulation of experience, and in particular the measurement of temperatures, can be expected to show the need or otherwise for the calculation of stresses from the more precise temperature distributions.

Future Work

To check the assumptions and reasoning of the theory, some laboratory tests are to be made on the effect of local heating on the residual stresses of circumferential butt welds. The results of these tests will be reported later.

Acknowledgment

The author is glad to acknowledge the guidance and suggestions of the Assistant Director, Dr. A. A. Wells, in deriving this solution of the problem.

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B.W.R.A. REPORT

Some Aspects of Cracking in Welded Cr-Ni Austenitic Steels

A SURVEY OF THE LITERATURE

By J. C. Borland, A.I.M., *and R. N. Younger*, M.A. (*Cantab.*)

Introduction

THE elevated temperature properties and excellent corrosion resistance in many acid media of the Cr–Ni austenitic steels, commend them for many uses.[46, 102, 134] Of particular interest at the present time is the performance of this class of steels in high-efficiency power stations.[124, 142, 144, 149–151]

As far as welding is concerned these steels can be broadly divided into two groups; those which are fully austenitic and those which contain some free (δ) ferrite. Whether or not the materials are partially ferritic depends upon the ratio of austenizers (*e.g.*, C, Ni, Mn, Cu, Co) to ferritizers (*e.g.*, Al, V, Ti, Si, Nb, Mo, Ta, W, Cr) present and the thermal (including welding) and mechanical treatment the material has undergone since being cast.

At elevated temperatures, the stress-to-rupture properties are reduced when ferrite is present and "sigma" phase,[109] which is hard and non-ductile, may form at temperatures between about 565° and 925°C. The presence of ferrite in mildly corrosive liquids is considered to be only a minor factor, but in severe corrosive conditions the ferrite may cause a local electrolytic cell action that can be most damaging.

Grain boundary attack, which is a particular form of corrosion, may occur owing to the depletion of chromium adjacent to the crystal boundaries. This loss in chromium is known to be due to the action of carbon, which forms a carbide ($Cr_{23}C_6$) containing 94% chromium.[83] Some effort has been expended in reducing the carbon content of stainless steels (down to 0·02–0·04%) to make them more resistant to attack, but an alternative remedy is to add 'stabilizers',[26, 67, 69, 70, 95, 143] which to some extent tie up the carbon and thus prevent it uniting with the chromium. In the U.K. titanium has been preferred as a 'stabilizer' while in the U.S.A. and in the U.S.S.R. niobium is used extensively. The amount and type of carbide formed is dependent on steel melting and rolling practice, and on subsequent heat-treatment procedures used.[93]

Since carbon is an austenite forming element, a secondary effect of adding carbide formers is to promote the formation of ferritic structures. Molybdenum, which is also considered to be a ferrite promoter, is added because of the increased corrosion resistance it confers on material subjected to attack by high concentrations of sulphuric acid and some chloride solutions. Such solutions would cause severe pitting attack on 18 Cr–8 Ni–Ti type steel. The addition of silicon or tungsten has a beneficial effect on the resistance to scaling. Sulphur and phosphorus form low-melting-point compounds with iron and nickel and therefore are undesirable, but unfortunately they cannot be eliminated altogether in manufacture. The effects of sulphur may be counteracted by the addition of manganese.

Normally the austenitic steels (Table I) as a whole have excellent weldability provided that the proper practices and procedures[20, 21, 74, 83, 85, 88, 101, 137] are employed. Special precautions must be taken when welding the free-machining grades (Type 303) otherwise hot cracking and porosity can occur. This is due to the high (relatively) proportions of phosphorus and selenium, or sulphur present. Here, dilution is an important factor to be considered, as it is also when welding the high silicon grades (Types 302B and 314). The stabilized grades (Types 321 and 347) are prone to heat-affected zone (underbead) cracking in heavily restrained sections, and as yet no completely satisfactory explanation of the cause has been suggested. Type 347 material is used to a large extent in steam

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Mr. Borland, formerly Head of the Reactive Metals Group at the Association, is now with the U.K.A.E.A., Culcheth. Mr. Younger is a Research Officer at the Association's laboratories

Table I

Standard and tentative standard types of wrought stainless and heat-resisting steels: chemical ranges and limits AISI austenitic types only

Note: Single values shown are maximum percentages except where otherwise noted

		Chemical Composition, %							
Type No.	*Composition*	*C*	*Si*	*Mn*	*P*	*S*	*Cr*	*Ni*	*Other Elements*
201	17–4–6	0·15	1·00	5·5	0·060	0·030	16·0 –18·0	3·5 – 5·5	N 0·25 max
202	18–5–8	0·15	1·00	7·5–10·0	0·060	0·030	17·0 –19·0	4·0 – 6·0	N 0·25 max
301	17–7	0·08/0·20	1·00	2·00	0·045	0·030	16·00–18·00	6·00– 8·00	.....
302	18–8	0·08/0·20	1·00	2·00	0·045	0·030	17·00–19·00	8·00–10·00	.....
302B	18–8 Si	0·08/0·20	2·00–3·00	2·00	0·045	0·030	17·00–19·00	8·00–10·00	.....
303	18–8 FM	0·15	1·00	2·00	See Note 2		17·00–19·00	8·00–10·00	Note 2
304	18–8	0·08	1·00	2·00	0·045	0·030	18·00–20·00	8·00–11·00	.....
304L	18–8 (LC)	0·03	1·00	2·00	0·045	0·030	18·00–20·00	8·00–11·00	.....
305	18–11 FS	0·12	1·00	2·00	0·045	0·030	17·00–19·00	10·00–13·00	.....
308	20–10	0·08	1·00	2·00	0·045	0·030	19·00–21·00	10·00–12·00	.....
309	25–12	0·20	1·00	2·00	0·045	0·030	22·00–24·00	12·00–15·00	.....
309S	25–12 (LC)	0·08	1·00	2·00	0·045	0·030	22·00–24·00	12·00–15·00	.....
310	25–20	0·25	1·50	2·00	0·045	0·030	24·00–26·00	19·00–22·00	.....
310S	25–20 (LC)	0·08	1·50	2·00	0·045	0·030	24·00–26·00	19·00–22·00	.....
314	25–20 Si	0·25	1·50–3·00	2·00	0·045	0·030	23·00–26·00	19·00–22·00	.....
316	18–12 Mo	0·10	1·00	2·00	0·045	0·030	16·00–18·00	10·00–14·00	Mo 2·00–3·00
TS 316	18–12 Mo	0·10	1·00	2·00	0·045	0·030	16·00–18·00	10·00–14·00	Mo 1·75–2·50
316L	18–12 Mo (LC)	0·03	1·00	2·00	0·045	0·030	16·00–18·00	10·00–14·00	Mo 1·75–2·50
317	19–12 Mo	0·10	1·00	2·00	0·045	0·030	18·00–20·00	11·00–14·00	Mo 3·00–4·00
321	18–10 Ti	0·08	1·00	2·00	0·045	0·030	17·00–19·00	8·00–11·00	Ti 5 × C min
347	18–10 Nb	0·08	1·00	2·00	0·045	0·030	17·00–19·00	9·00–12·00	Nb 10 × C min
TS 347	18–10 Nb	0·08	1·00	2·00	0·045	0·030	17·00–19·00	9·00–12·00	Nb 8 × C min
TS 347A	18–10 Nb Ta	0·08	1·00	2·00	0·045	0·030	17·00–19·00	9·00–12·00	Nb-Ta 8 × C min

NOTE 1. Prefix TS indicates type number is a tentative standard.
NOTE 2. Phosphorus or sulphur or selenium, 0·07% min. zirconium or molybdenum 0·60% max.
NOTE 3. The maximum content of molybdenum of 2·50% in 316L and TS 316 is established in accordance with National Production Authority Order M–52 Mar. 31, 1951.
NOTE 4. The minimum ratio of niobium to carbon in TS 347 of not greater than 8 to 1 and the minimum ratio of niobium tantalum to carbon in TS 347A of not greater than 8 to 1 are established in accordance with National Production Authority Order M–3 as amended Mar. 15, 1951.

piping applications, and in recent years serious cracking troubles have been experienced under service conditions. It is believed that these troubles can be minimized by using a modified Type 316 electrode (16 Cr–8 Ni–2 Mo).[156]

It would appear that most weld-metal cracks form during or immediately after welding, and are probably due to adverse strains imposed upon the weld joint while it is cooling. Most cracks originate while the weld is semi-solid, but others, it is conjectured, may form during subsequent cooling from the solidus. Cracking can also occur during thermal stress-relieving treatment or during service. Relief from cracking troubles during welding may be gained by additions to the weld, to ensure that the deposit, on cooling to room temperature, contains a small proportion (5–8%) of the ferrite phase. Welding electrodes developed specifically for this purpose have been available for some time.[97] The presence of ferrite is, however, undesirable in many cases because of sigma formation during service at elevated temperatures.

Base metal cracks can also form during or immediately after welding. Cracking during welding is thought to be due to liquation of the grain boundaries (over-heating) or to some form of high-temperature brittleness. Cracks forming after welding, particularly during stress relieving or service are probably stress-rupture type failures.

There is no doubt that the weld and base metal composition, restraint, section thickness, and joint geometry are all important factors affecting the cracking tendency of Cr–Ni austenitic weldments. Generally, weldments in thin gauge material are not crack sensitive. Increasing the section thickness increases the cracking susceptibility.

The considerable literature on this subject has been reviewed and conclusions have been drawn which may help to provide a better understanding of the basic mechanisms—metallurgical and mechanical—causing cracking in these steels.

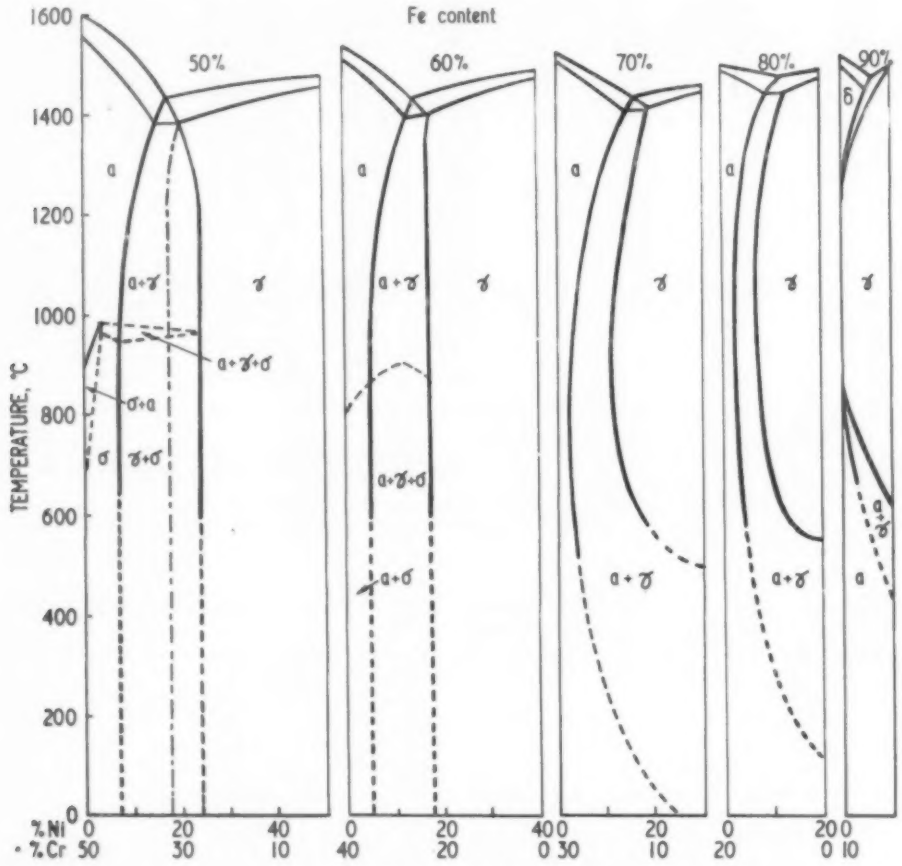
CONSTITUTION OF CHROMIUM–NICKEL AUSTENITIC STEEL

Iron–Chromium–Nickel Ternary System

The first comprehensive investigation of this system was in 1927 by Bain and Griffiths.[2] They established the positions of the alpha (α), gamma (γ), and alpha plus gamma ($\alpha+\gamma$) phase fields on the iron side of the diagram, and also noted the existence of a brittle constituent which they called "B". This constituent is now known as 'sigma'.[40, 68] Further notable studies were conducted by Wever and Jellinghause,[3] Jenkins, Bucknall, Austin and Mellor,[5] and Schafmeister and Ergang.[9] Recently, Pugh and Nisbet[56] re-examined the work of the several authors and compared the collective results with data they had obtained in their own laboratories. In general, good agreement was noted, except perhaps for the vertical section constructed at 50% iron.

Vertical sections at 50–90% iron, which are of particular interest to the weld cracking problem are shown in Fig. 1.

A stainless steel containing 70% iron, 16% chromium and 14% nickel, it will be noted from Fig. 1, will initially solidify as alpha and gamma iron. On the other hand, a stainless steel containing about 15% chromium and 35% nickel on cooling will solidify



1—*Vertical sections of Fe–Cr–Ni system*[56]

only as gamma iron. The different thermal histories of these two alloys must therefore be considered in interpreting the effects of elements on their respective weld-cracking tendencies. In view of the wide differences in the solubilities of elements in alpha and in gamma iron, the high temperature as well as the room temperature microstructural features, particularly in the 18/8 class of steels, should be of significance in relation to the weld cracking problem.

Maximum Solubilities of Elements in Alpha and Gamma Iron

Since a large number of hot cracks are associated with liquid films it is important to compare the maximum solubility of elements in delta and gamma iron. Thus, it is useful to compare the relative solubilities of elements in pure iron binary systems. This information is shown in Table II and was compiled from data by Hansen and Anderko.[159] Elements that are soluble in gamma iron up to 3% (by weight) are: Be; B; C; Al; Si; P; S; Ti; V; Zr; Nb; Sn; Sb; and Ta. Elements soluble in delta iron to less than 3% are: C; P; S; and Mn.

Room Temperature Microstructure

As previously mentioned, many of the Cr–Ni stainless steels[20, 52, 77, 101, 137] are not fully austenitic (particularly the 18/8 type) and contain small amounts of ferrite.[91, 111, 112, 116, 131, 160] The actual phases present depend on: (*a*) composition; (*b*) heat-treatment, and (*c*) cold working. Castings and weld deposits are

Table II

Solubility of elements in delta and gamma iron

Element	*Solubility in delta, wt %*	*Temp., °C.*	*Solubility in gamma, wt %*	*Temp., °C.*
Li	insoluble		insoluble	
Be	~7·4	1165	~0·5	~1050 loop
B		1381	0·021	1149
C	0·10	1493	2·06	1149
Na	insoluble		insoluble	
Mg	insoluble		insoluble	
Al	36	1232	0·6	~1150 loop
Si	18·5	1200	2·15	~1170 loop
P	2·8	1050	0·25	1150 loop
S	0·18	1365	0·05	1356
K	insoluble		insoluble	
Ca	insoluble		insoluble	
Sc				
Ti	~7	1340	0·6	1150 loop
V	continuous		~1·5	1150 loop
Cr	continuous		12·5	~1050 loop
Mn	1	1504	continuous	
Co	17·3	1499	continuous	
Ni	3·4	1512	continuous	
Cu	~6	1484	8·5	1094
Zn			46	782
Ga	?		?	
Ge		1125	2·5	1150 loop
As	11	835	3·75	1150 loop
Se				
Rb	insoluble		insoluble	
Sr	insoluble		insoluble	
Y	?		?	
Zr	~7	1335	<0·15	1335
Nb	4·5	1360	2·0	1220
Mo	37·5	1450	2·7	1100 loop
Tc	?		?	
Ru	?		?	
Rh	?		?	
Pd	?		?	
Ag	?		?	
Cd	insoluble		insoluble	
In	?		?	
Sn	~16·1	1130	2–215	1150 loop
Sb	~35	1002	~2·0	1150 loop
Te	?			
Cs	insoluble			
Ba	insoluble			
La				
Ce	very slight solubility	1390	very slight solubility	
HF	?		?	
Ta	~7	~1410	3·0	1220
W	33	1540	3·2	1150 loop
Re	?		?	
Os	?		?	
Ir	?		?	
Pt	?		?	
Au	?		?	
Hg	?		?	
Tl	?		?	
Pb	?		?	
Bi	?		?	

strongly affected by solidification rates, and thus the proportions of alpha and gamma iron will not necessarily be similar to that of the wrought material. It is fortunate, however, that wide variations in cooling rate are not significant.

The effect of alloying elements on the structures found in wrought quench-annealed, chromium-nickel steels have been studied successively by Strauss and Maurer,[1] Scherer, Riedrich and Hoch,[10] and by Schaeffler[37] who proposed a modified structural diagram for stainless steel. This is shown in Fig. 2. An empirical formula to indicate the microstructure, developed by Newell and Fleischman[6] and later modified by Post and Eberly[59] takes the form:

$$\text{Ni value} = \frac{(\text{Cr}+1{\cdot}5\,\text{Mo}-20)^2}{12} - 0{\cdot}5\,\text{Mn} - 35\,\text{C} + 15 \quad \ldots \quad (1)$$

and was proposed to account for the effect of various elements on the structural constituents developed. The homogenized material was fully austenitic if the nickel content of the weld metal exceeded the nickel value given by the sum of the elements on the right hand side of the equation. An alternative formula[6] modified by Field, Bloom and Linnert[22] is as follows:

$$\text{Ni value} = \frac{(\text{Cr}+2\,\text{Mo}-16)^2}{12} - 0{\cdot}5\,\text{Mn} + 30\,(0{\cdot}10-\text{C}) + 11 \quad \ldots (2)$$

Because of non-equilibrium freezing conditions in castings and welds, the effects of various elements are modified, and thus several investigators have suggested alterations in their values.[6, 12, 22, 29, 49, 51, 59] For welds a widely used formula is:

$$\text{Ni value} = 1{\cdot}1\,(\text{Cr}+\text{Mo}+1{\cdot}5\,\text{Si}+0{\cdot}5\,\text{Nb}) - 0{\cdot}5\,\text{Mn} - 30\,\text{C} - 8{\cdot}2 \quad \ldots\ldots\ldots\ldots \quad (3)$$

The structural diagram for stainless steel weld metal developed by Schaeffler is based on this formula. The Schaeffler diagram,[50] as it is commonly called, is shown in Fig. 3. The nil ferrite limit[126] for wrought material is superimposed for completeness. The diagram has been used widely, and apparently successfully, for predicting the percentage ferrite (to within $\pm 4\%$) in weld metal deposits.

Taking into account the neglected effect of nitrogen, which is a strong austenizer, De Long, Ostram and Szumachowski[131] have suggested an alteration to the Schaeffler diagram in the region 12–24% Ni, 18–26% Cr. This area is shown in Fig. 4. The proposed nickel equivalent for nitrogen is 30. The authors claim that the modified diagram should be useful in controlling the ferrite content of weld metal within $\pm 2\%$. This is better than the $\pm 4\%$ tolerance claimed by Schaeffler for his work.

NATURE AND INCIDENCE OF CRACKING

Cracking in the weld, and in the base metal adjoining it, is due to the additive effects of metallurgical and mechanical factors, and is similar in appearance and in origin to corresponding defects occurring in castings. 'Hot' and 'cold' cracks are both known to occur, and are so called to distinguish between the temperatures at which they form. 'Hot' cracking is usually associated with hot tearing and is generally believed to occur while the metal solidifies. 'Cold' cracking, which occurs at relatively low temperatures, is developed when the metal is solid and contains no liquid films. All hot cracks are intergranular.

Certain forms of weld-metal crack are distinctive in appearance, and have been appropriately named. Crater cracks, for instance, are situated in the weld crater and occur when the welding operation is completed or interrupted.[48] This form of defect is generally peculiar to arc welding, and is most troublesome when

depositing the root pass of a highly restrained joint of the double-V type.[48] The frequency of these cracks appears to decrease with increasing ferrite content and decreasing niobium content.[98]

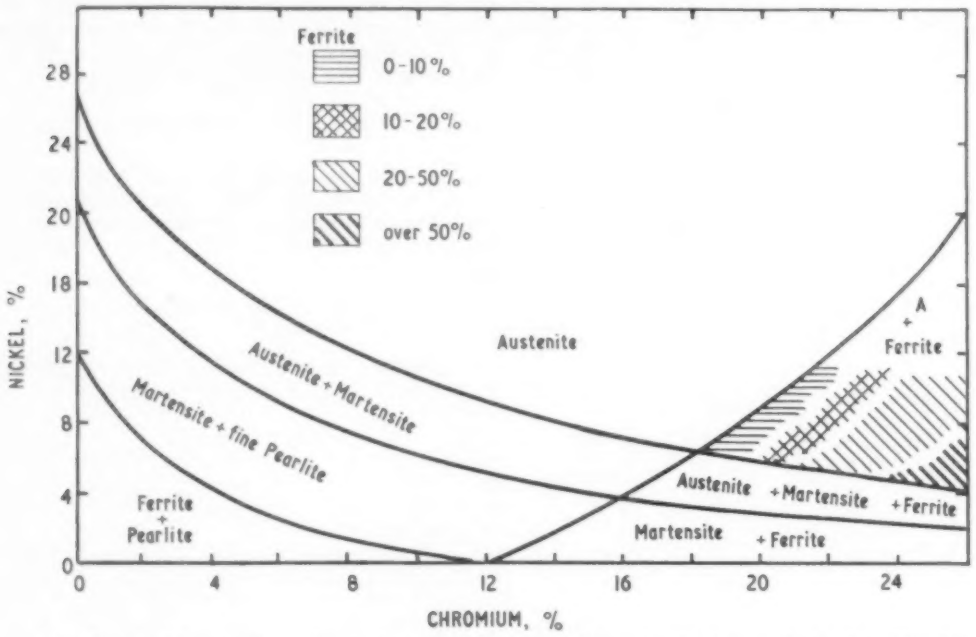
Root (notch) cracks are also prevalent in highly restrained double-V joints, and additionally often occur when a backing strip or ring is used. It is generally believed[98] that the cause of the trouble is unsatisfactory joint design or welding technique, or both.

Star cracks may be described as fine cracks generally $\frac{1}{16}$–$\frac{3}{32}$ in. long, radiating from a central point; their similarity in appearance to crater cracks suggests that they might be due to similar causes, particularly in regard to the ferrite and niobium contents.[98] They can also be formed alongside the weld by striking the arc momentarily on the base metal. Star cracks are also caused by splash droplets[76] but they are not generally observed until after the droplet is removed. One fabricator, according to a survey carried out in America,[98] thought they were caused by welding over craters which had not been removed before depositing the next layer of weld metal.

Longitudinal cracks run in the direction of welding, and as they most often occur along the centreline of the weld it is not unreasonable to suggest that they occur while the weld is still very hot, or even before it completely solidifies. Most longitudinal cracks, since they are a severe form of defect, are generally seen with the unaided eye on the top surface of the weld. They are most common in the first pass of a multipass weld.[48]

Transverse cracks lie perpendicular to the fusion boundaries, and are caused by excessive longitudinal stresses. These cracks may extend beyond the weld bead into the base metal. They may originate at the fusion boundary, when they are designated fusion-line cracks. These are thought by some workers[103] to be caused by low-melting-point segregates which, on melting, allow a notch type defect to be formed; the incipient crack can then propagate, because of relatively high longitudinal stresses being imposed at this moment.

Micro-cracking and fissuring are terms usually restricted to describing small cracks which occur in the weld metal in a random direction, and which are frequently not seen except by metallographic examination. This form of cracking is very difficult to avoid in fully austenitic weld metals, and as yet no completely satisfactory explanation of it has been given.



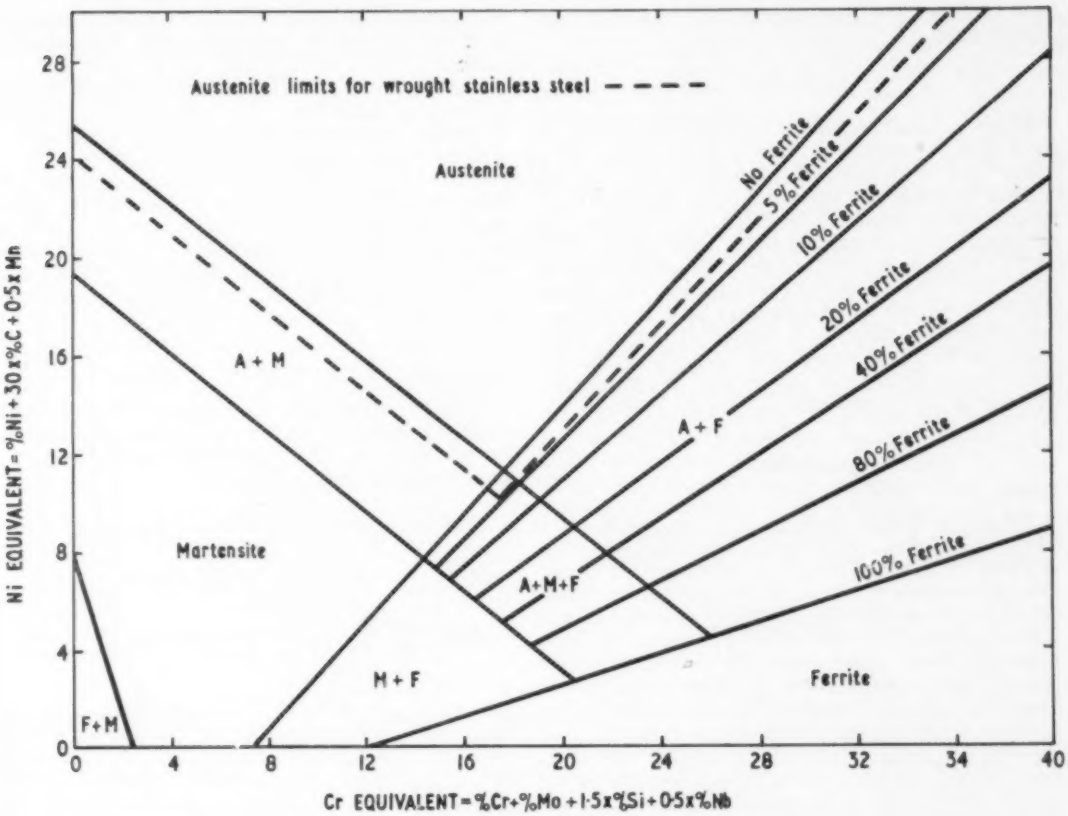
2—*Structural diagram of wrought quench-annealed chromium nickel steels, after Maurer. Region from 17 to 26% Cr and 3 to 11% Ni verified by Scherer–Riedrich–Hoch for 0·07–0·13% C, 0·30–0·48% Mn and 0·23–0·37% Si (after Schaeffler)*[37]

Base-metal cracking, though not so common as weld-metal cracking, is particularly frequent in Type 347 (18 Cr–13 Ni–1 Nb) material, and occurs most often in sections whose thickness exceeds $\frac{3}{4}$ in.[98] In general, the cracks do not extend to the weld metal. The types of base-metal crack are referred to here as:

(*a*) Longitudinal base-metal cracks—cracks parallel to the fusion line
(*b*) Fusion-line cracks—cracks at right angles to the fusion line and in contact with it
(*c*) Weld/base-metal cracks—cracks at right angles to the fusion line and continuous from base metal to weld metal.

The causes of these latter cracks are unknown, although some suggestions have been advanced. One investigator,[147] for instance, suggests that 'surface active' elements in the weld penetrate into the grain boundaries of the base metal and lower the intergranular strength, thereby causing low-stress fractures.

A survey has been carried out by Poole[98] in America,



3—*Constitution diagram for stainless steel weld metal*[60]

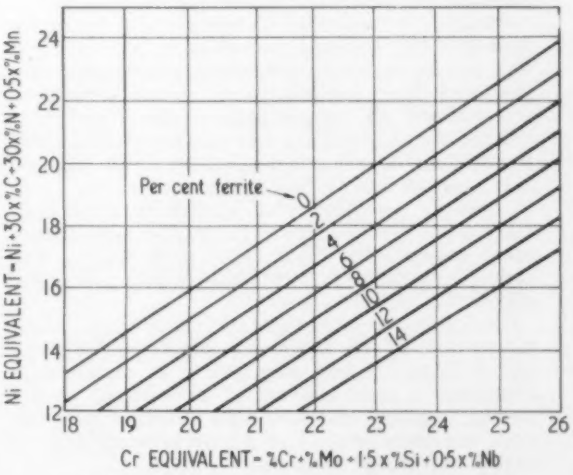
among fabricators of Type 347 steel, and it is reported that 100% of them experienced crater-cracking troubles, 78% weld cracking and microfissuring, and 39% base-metal troubles.

A similar survey has not yet been carried out in the U.K.

FACTORS AFFECTING WELD-METAL CRACKING

Weld-metal cracks occur when the fracture strength is exceeded by the stresses developed during welding. These stresses are created by thermal expansion brought about by the fusion process and subsequent contraction strains. They are mainly a function of joint design, welding technique,[18, 30, 48, 55, 58, 64, 71, 88] jigging arrangements[88] and the weld thermal cycle. Since austenitic stainless steels possess a high thermal expansion coefficient the weld shrinkage strains if unhindered are high. The resultant shrinkage stresses may thus reach the yield strength of the material.[140] The ductility of the weld in uniaxial tension is rarely exceeded, and so to explain cracking many investigators[11, 15, 16, 23, 38] have suggested that multiaxial stresses are responsible. A weld made under complete multiaxial restraint would either crack or leave some form of shrinkage cavity.

Several weld cracking tests[87] have been devised to assess the cracking susceptibilities of materials. Unfortunately however, these tests, although being useful in assessing relative cracking tendencies, do not allow direct comparisons to be made between themselves because of the widely different conditions of external restraint imposed upon the fused zone. Where possible in the following pages, the present authors



4—*Effect of nitrogen on the Schaeffler diagram*[131]

have indicated the type of test used by the various authors cited.

Preheating

Being non-hardenable on cooling, preheating of these steels is generally considered unnecessary, and is indeed not recommended because of the reduced cooling rate through the 'sensitizing' range.[114,158] A mild preheat, possibly 300–500°F. (150–250°C.), is sometimes used in welding heavy wall tubing (Type 347) to improve the uniformity of penetration of the root pass, and to prevent entrapment of welding slag.[158] Thielsch[61] comments that preheating heavy sections of Type 347 steel is necessary to prevent weld cracking. He omits however to give further details. Weld shrinkage stresses are reduced by preheating. This fact is often overlooked, yet it may be important.

Delta Ferrite

General

The action of delta ferrite in reducing, and in many cases preventing, weld-metal cracking was first reported by Riedrich and his co-workers.[10,14] Generally, the presence of more than 5% will prevent cracking in Cr–Ni weldments. Thomas[28] believed that its beneficial action was due to its presence in a finely dispersed form, which prevented the formation of fissures or the propagation of them once formed. Linnert and Bloom[87] suggested that the ferrite reduced cracking by increasing the boundary area, and thus reducing the effective width of the grain boundary. In contrast to these theories, Lyubavskii and Toropov[135] and Parks[124] advanced the opinion that the favourable effect of ferrite was due to its capacity for dissolving larger quantities of certain elements, particularly phosphorus,[124,135] silicon,[124,135] and niobium,[135] than austenite. This effect reduced segregation, and thereby strengthened the grain boundary. Rollason and Bystram[76] expressed a similar viewpoint, at least in regard to silicon, but in addition suggested that the greater grain boundary area, and lack of coincidence of final grain boundaries and primary coring were contributory factors in preventing weld cracking.

Medovar,[94] in addition to agreeing that ferrite was beneficial by dispersing and thinning out the liquid film, and also by lessening the tendency for segregation, suggested that ferrite-forming elements probably strengthened the intergranular layers, and thus reduced the danger of hot cracking. On this last point, however, his meaning is not entirely clear. Medovar also suggests that ferrite-formers such as Al, Ti, V, Si, and Cr combined with sulphur to form high-melting-point compounds and thus offset low-melting-point segregations which could form cracks. This action, he noted, was particularly beneficial in highly alloyed Cr–Ni welded joints.

The theory that segregation, particularly micro-segregation, is very harmful is supported by Noren and Hellestein,[61] who suggested that compositions containing a small amount of delta ferrite gave a smaller freezing range than fully austenitic compositions and these resulted in a reduced segregation of oxygen and 'other' elements.

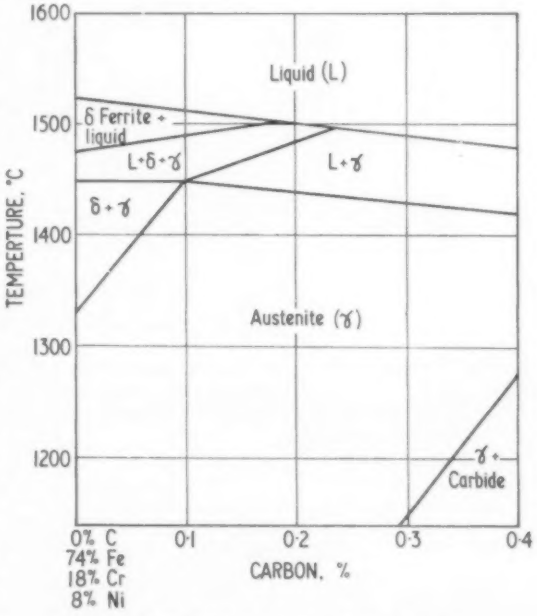
Heuschkel[125] attributes the lesser cracking tendency with ferrite to the lowered strength and improved hot-ductility at 1800°F. (982°C.) when this phase is present.

Lashko and Lashko-Avakyan[157] suggest that the influence of delta ferrite in reducing hot cracking is primarily connected with the appreciable reduction in shrinkage strains when delta ferrite is present.

The various theories that have been proposed are not necessarily mutually exclusive, and all the factors cited as being beneficial probably contribute towards the prevention of cracking. There is, however, general agreement that weld compositions producing 5% or more ferrite in the room-temperature structure are not susceptible to weld metal cracking.

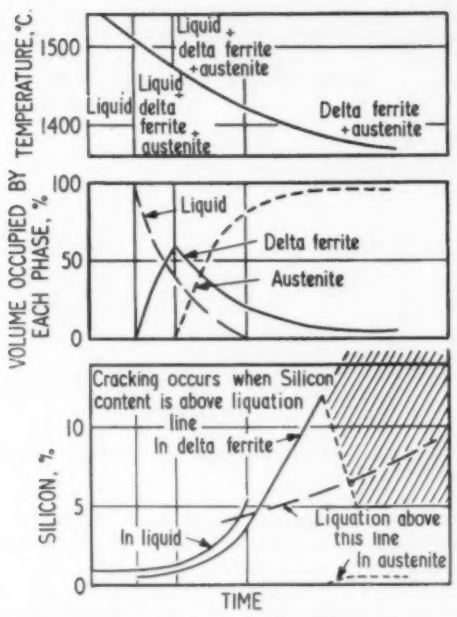
Theory of Parks

Parks[124] considers that ferrite may be beneficial, but in some instances it can also act to cause hot cracking by redistributing and segregating harmful elements, such as phosphorus and silicon.



5—The 18 Cr–8 Ni stainless steel phase diagram

Taking the example of 18/8 weld metal containing 0·10% C (phase diagram Fig. 5), he explains that, as the molten metal solidifies, ferrite is the first phase to form and thus selectively absorbs the ferrite formers. On cooling to the peritectic temperature, the residual melt decomposes into austenite and ferrite, or reacts with ferrite to form austenite, depending on the carbon content. On further cooling, the ferrite is progressively absorbed by the continuing growth of austenite, and because of the difficulty of accommodating the ferrite-formers in the austenitic lattice, they tend to accumulate at the instantaneous austenite/ferrite interface. When the accumulation of elements exceeds a critical amount, liquation of the grain boundaries occurs, and if sufficient stresses are



6—*Mechanism of hot cracking in 18/8 stainless steel considering Si as the cause of delta ferrite liquation*[124]

imposed at this moment, there is hot cracking. Phosphorus and silicon are suggested as being mainly responsible for hot cracking. The amount of ferrite required to prevent hot cracking is determined by the phosphorus and silicon contents of the weld metal, and the maximum amount of ferrite which forms at the peritectic temperature. A pictorial illustration of what happens when the weld solidifies is shown in Fig. 6. Parks claims that increasing the carbon content is beneficial because it decreases the amount of ferrite formed as the weld solidifies, and thus lessens the effect of ferrite in concentrating and segregating the injurious elements. The following example is quoted. "Consider two stainless steel weld puddles both containing 0·75% Si, one with 0·10% C, the other with 0·20% C, and both balanced to give 5% residual ferrite. The one with 0·10% C will form a maximum of, say, 70% delta ferrite, whereas a 0·20%-C steel will form a maximum of, say, 15% delta ferrite. At the time of maximum ferrite formation, if the Si content is 0·70% and if none is lost as the volume of ferrite contracts, the final Si in the two ferrites will be 9·8 and 2·1% respectively. The first steel should 'hot crack' in the weld metal and the second should not."

Two criticisms may be made of this theory: (1) Under weld cooling conditions a justifiable assumption would be that the room-temperature ferrite content is related to the amount of ferrite existing at high temperatures, increasing ferrite at room temperature indicating increasing ferrite content at elevated temperatures (or vice versa). With Parks' mechanism, increasing the ferrite content at high temperatures is detrimental. This is not consistent with the view that higher room-temperature ferrite contents are beneficial; (2) carbon is a strong austenizer, and thus increasing the carbon content should not only ensure less ferrite in the microstructure but may even make the weld metal fully austenitic at all temperatures.

Chemical Composition

Sulphur

Campbell and Thomas[29] found that the addition of sulphur (0·006–0·044%) to fully austenitic 25 Cr–20 Ni weld metal containing approximately 0·15% C and 1·6% Mn had an adverse effect on tensile ductility. These authors say that specimens failing at low ductilities exhibited numerous fissures; but they fail to state which particular specimens were defective. In a discussion of their paper Linnert and Bloom[29] confirmed that the ductility decreases with increase in sulphur content. From data given by these authors (Fig. 7) it can be noted that the tensile strength remained sensibly constant up to 0·03–0·04% S and thereafter it decreased. It seems likely therefore that these materials were susceptible to cracking at sulphur contents in excess of about 0·03%.

Table III

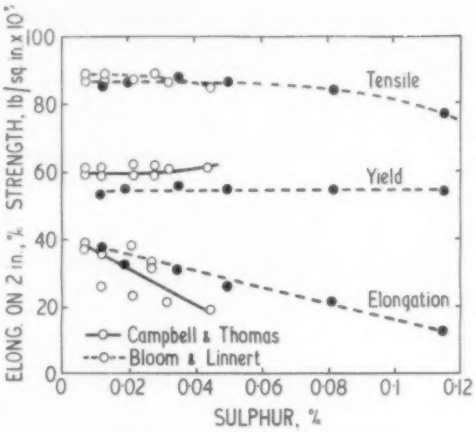
Effect of sulphur and phosphorus on 15 Cr–35 Ni weld metal properties[45]

Sample No.	*Composition, %* C	Si	Mn	S	P	Cr	Ni	*Yield strength, lb/sq.in.*	*Tensile strength, lb/sq.in.*	*Elong., %*	*Red. in area, %*	*Observations*
Sulphur Variation												
4908	0·256	0·42	1·71	0·010	0·012	15·22	33·91	58,500	91,500	37	47·5	Sound
4910	0·232	0·42	—	0·017	—	—	—	58,500	91,500	36	49	Sound
4911	0·244	0·42	—	0·019	—	—	—	58,000	91,000	34	40·5	Sound
4912	0·236	0·42	1·66	0·023	0·010	15·22	33·42	60,000	93,000	38	45	Sound
4967	0·220	0·42	—	0·035	—	—	—	57,000	86,000	22	22	Few fissures, porosity
Phosphorus Variation												
4908	0·256	0·42	1·71	0·010	0·012	15·22	33·91	58,500	91,500	37	47·5	Sound
4816	0·225	0·49	1·67	0·009	0·016	15·46	33·35	59,500	91,000	38	50·5	Sound
4817	0·240	0·54	—	—	0·028	—	—	57,500	83,500	22	30·5	Many fissures
4818	0·207	0·55	—	—	0·030	—	—	55,500	79,500	19·5	23·5	Many fissures
4819	0·222	0·50	—	—	0·032	—	—	55,000	69,500	11	17	Many fissures
4820	0·218	0·53	1·76	0·010	0·038	15·31	33·60	56,000	81,500	21·5	22	Many fissures
4820R								54,500	75,000	15	20·6	Many fissures slag
4863	0·227	0·48	1·78	0·009	0·042	15·32	—	52,000	59,500	8	17	Many fissures slag

The effect of sulphur on the tendency to cracking of 25 Cr–20 Ni weld metal was also studied by Medovar,[105] who found that the addition of up to 0·5% S caused large hot cracks in welded 0·47 in. thick T-shaped specimens. Adding about 6% S eliminated cracking, due to a refinement of the microstructure, together with the large quantity of a complex Fe–Ni–S eutectic. On vacuum heating the sample containing 6% S, for 30 min at a temperature of 1000–1100°C., the eutectic melted out. It was also noted that drops of eutectic which emerged from the weld impinged upon and diffused deep into the adjoining base metal. Grain boundaries were particularly affected.

Rozet *et al*[45] investigated the effect of sulphur on fully austenitic 15 Cr–35 Ni weld metal (Type 330) and the results which they tabulate (Table III) show that 0·023% S did not affect the soundness of the weld deposit, but increasing the sulphur content to 0·035% caused fissures and porosity. Carbon and manganese contents were approx. 0·24% and 1·66% respectively.

A study of the inter-related effects of sulphur and manganese on fully austenitic weld metal containing basically 15 Cr–25 Ni–6 Mo (0·14% C, 0·18% Si) was made by Lyubavskii and Toropov.[135] These authors found that 0·036% S caused hot cracks when the manganese content was 1·0%, but increasing the Mn to 2·0% prevented cracking until the sulphur was further increased to 0·052% (Table IV). The same authors made a similar investigation with 16 Cr–13 Ni–2 Mo (0·09–0·13% C, 0·10–0·28% Si) material which was fully austenitic only at room temperature. They reported that, for a low Mn content (0·20–0·28%), a sulphur content of 0·026% caused cracking, but 0·034% did not when the Mn was increased to 1·00–1·16%. A high Mn content (2·14–2·34%) permitted the sulphur content to be raised above 0·054% (Table V).



7—Comparison of data on the effect of sulphur in 25 Cr–20 Ni weld metal[29]

Toropov[130] investigated the effect of sulphur additions to 15 Cr–25 Ni–6 Mo steel containing high carbon and silicon contents (0·74–0·80% C, 0·80–0·92% Si, 0·79–0·96% Mn). Cracks were observed at 0·044% and 0·074% S but not at 0·028% (Table VI).

Kauhausen and Vogels[115] report that core wire for electrodes to be used in welding 16 Cr–13 Ni material can tolerate no more than 0·01% S if hot cracking is to be avoided. The carbon and manganese contents

Table IV

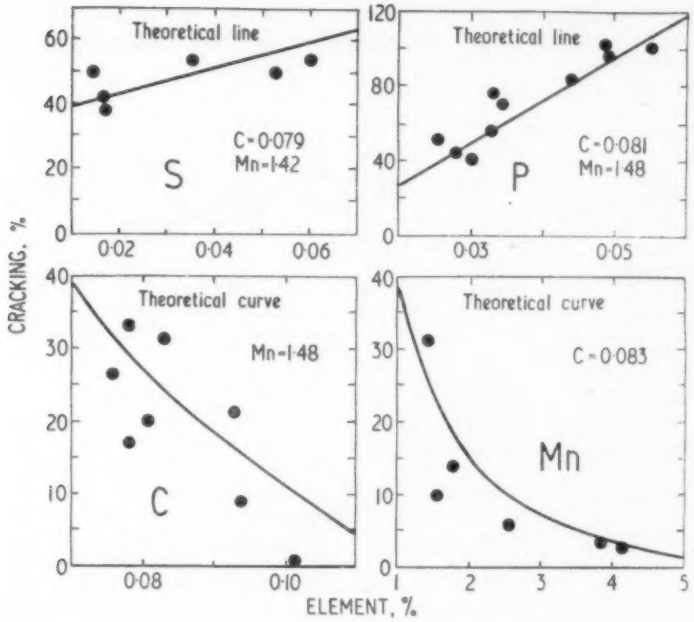
Influence of sulphur and manganese on hot cracking in 15 Cr–25 Ni–6 Mo welds[135]

Test Pieces	*Composition, %*								*Presence of Cracks*
	C	*Si*	*Mn*	*S*	*P*	*Cr*	*Ni*	*Mo*	
AS0	0·14	0·17	1·02	0·031	0·022	14·15	25·10	5·80	nil
AS1	0·13	0·18	1·01	0·036	0·021	14·35	25·22	5·90	cracks
AS2	0·17	0·20	1·01	0·080	0·019	14·60	25·20	5·90	cracks
AS3	0·15	0·19	1·08	0·110	0·020	13·70	25·54	5·85	cracks
AS4	0·13	0·17	1·96	0·032	0·028	14·09	24·05	5·70	nil
AS5	0·15	0·18	2·04	0·042	0·025	14·00	24·36	5·70	nil
AS6	0·14	0·17	2·02	0·052	0·024	13·92	24·05	5·60	cracks
AS7	0·14	0·17	2·00	0·061	0·025	13·95	24·20	5·55	cracks

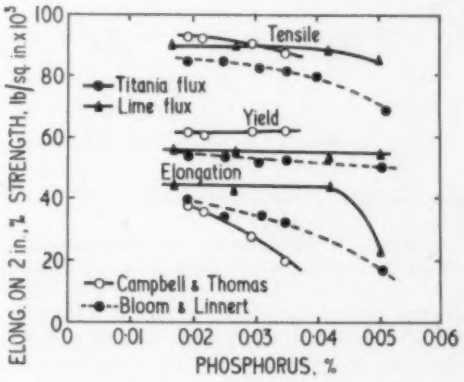
Table V

Effect of sulphur and manganese on hot cracking in 16 C–13 Ni–2 Mo welds[135]

Test Pieces	*Composition, %*								*Presence of Cracks*
	C	*Si*	*Mn*	*S*	*P*	*Cr*	*Ni*	*Mo*	
BS1	0·13	0·16	0·28	0·026	0·022	16·25	11·75	2·58	cracks
BS2	0·10	0·16	0·28	0·048	0·021	15·80	12·05	2·46	cracks
BS3	0·09	0·20	0·20	0·079	0·021	15·80	12·05	2·58	cracks
BS4	0·11	0·13	1·08	0·034	0·026	15·70	11·90	2·37	nil
BS5	0·11	0·12	1·16	0·045	0·027	15·85	11·70	2·35	cracks
BS6	0·09	0·16	1·00	0·092	0·025	15·85	12·26	2·47	cracks
BS7	0·10	0·10	1·02	0·130	0·024	15·80	12·19	2·47	cracks
BS8	0·12	0·28	2·34	0·040	0·032	15·70	12·42	2·52	nil
BS9	0·12	0·28	2·26	0·054	0·030	15·80	12·43	2·50	nil
BS10	0·11	0·24	2·22	0·071	0·030	15·70	12·43	2·55	cracks
BS11	0·11	0·28	2·14	0·092	0·031	15·70	12·03	2·49	cracks
BS12	0·11	0·20	2·26	0·110	0·033	15·60	12·43	2·43	cracks



8—*Effect of elements on segmented circular groove cracking*[152]



10—*Comparison of data on the effect of phosphorus in 25 Cr–20 Ni weld metal*[29]

reported in their paper were 0·10% and 1·2% respectively.

Adding 0·7–2·0% S to 18 Cr–8 Ni type weld metal (T-shaped tests) Medovar[105] found that cracking was absent, and the microstructure showed the sulphur eutectic formed in the shape of globules. No ferrite was present. Kazennov[99] (reported by Medovar[105]) noted that an increase in the sulphur content from 0·017–0·040% in a 18 Cr–8 Ni weld led to an almost fourfold increase in the susceptibility to hot cracking.

By varying sulphur additions, in the range 0·009 to 0·050%, to Type 347 metal (0·079% C, 0·32–0·73% Si, 0·42% Mn) Hoerl and Moore[152] show that for every 0·010% increase in the sulphur content the cracking increased by 3·8% (Fig. 8, segmented groove test).

Phosphorus

Campbell and Thomas[29] suggest that the effect of phosphorus on the tensile properties of 25 Cr–20 Ni weld metal (0·124–0·144% C, 0·37–0·85% Si) is interrelated with the effect of silicon. The phosphorus equivalent (P.E.) as these authors term it, is as follows:

$$\text{P.E.} = \text{P} + 0{\cdot}04\,(\text{Si} - 0{\cdot}25) \quad \ldots\ldots\ldots\ldots \quad (4)$$

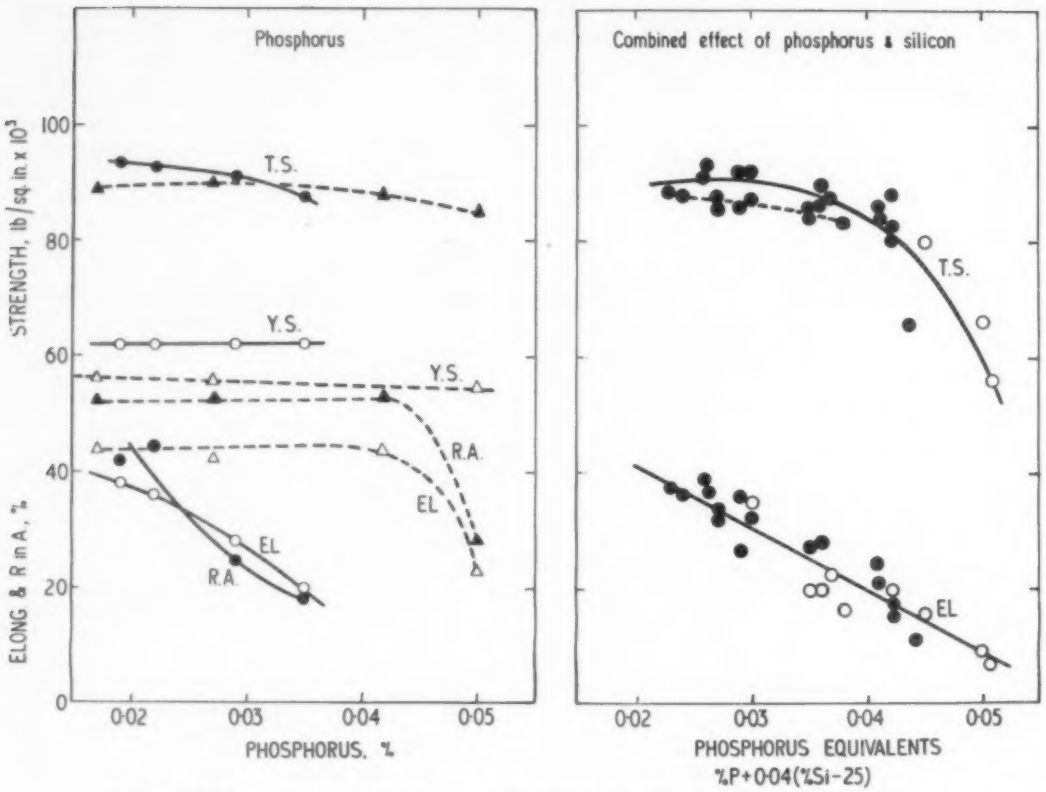
A graph of tensile strength against P.E. shows a characteristic falling off in strength beyond 0·030. Elongation versus P.E. approximates to a straight line of negative slope (Fig. 9). The effect of phosphorus itself is also shown in Fig. 9 and illustrates the differences afforded by different electrode coatings. At the same phosphorus levels micro-fissuring is less with the lime-type coated electrode than with the titania-type electrode, a point which is confirmed by Linnert and Bloom[29] (Fig. 10). The reasons for the better properties secured by the lime-type electrode were not satisfactorily explained, but it seems that the Si content of the weld metal deposited with the titania-coated electrode was possibly greater than that with the lime-electrode.

The influence of phosphorus on cracking and mechanical properties for 15 Cr–35 Ni weld metal (0·22–0·26% C, 0·42–0·55% Si) is shown in Table III. Weld metals containing 0·016% P and less were

Table VI

Effect of sulphur and phosphorous on weld cracking in 15 Cr–25 Ni–6 Mo welds[130]

Composition, %								
C	*Si*	Mn	*S*	*P*	*Cr*	*Ni*	*Mo*	*Cracks*
0·75	0·85	0·88	0·028	0·026	14·51	24·40	5·5	No
0·79	0·86	0·96	0·044	0·025	14·42	24·42	5·4	Yes
0·75	0·81	0·90	0·074	0·023	14·33	24·44	5·4	Yes
0·80	0·92	0·85	0·019	0·026	14·64	24·87	5·38	No
0·75	0·80	0·79	0·016	0·049	14·37	24·32	5·30	Yes
0·74	0·85	0·94	0·010	0·065	14·38	24·58	5·38	Yes



9—Effect of phosphorus and combined effect of phosphorus and silicon on 25 Cr–20 Ni weld metal[29]

sound, whilst welds with more than 0·028% contained fissures.[45]

The effect of phosphorus on the cracking susceptibility of 15 Cr–25 Ni–6 Mo (0·19% C, 0·15% Si) and 16 Cr–13 Ni–2 Mo (0·15% C, 0·27% Si) weld metal was studied by Lyubavskii and Toropov.[135] The 15 Cr–25 Ni–6 Mo weld metal was fully austenitic at all temperatures, whilst that of the 16 Cr–13 Ni–2 Mo was fully austenitic only at room temperature. With the 15 Cr–25 Ni–6 Mo welds, cracks were not formed in the presence of 0·028% P, but were formed when the phosphorus was increased to 0·035% (Table VII). In the 16 Cr–13 Ni–2 Mo alloy, cracking was absent with 0·036% but present with 0·044% P (Table VIII).

Table VII

Influence of phosphorus on hot cracking in 15 Cr–25 Ni–6Mo weld metal[135]

Test Pieces	*Composition, %*								*Presence of Cracks*
	C	*Si*	*Mn*	*S*	*P*	*Cr*	*Ni*	*Mo*	
AR1	0·14	0·16	1·05	0·015	0·028	14·60	24·51	5·40	nil
AR2	0·14	0·15	0·96	0·015	0·035	14·25	25·25	5·60	cracks
AR3	0·15	0·15	0·96	0·016	0·045	14·05	24·99	5·68	cracks
AR4	0·14	0·16	0·95	0·010	0·059	13·85	23·98	5·82	cracks

Table VIII

Influence of phosphorus on hot cracking in 16 Cr–13 Ni–2Mo weld metal[135]

Test Pieces	*Composition, %*								*Presence of Cracks*
	C	*Si*	*Mn*	*S*	*P*	*Cr*	*Ni*	*Mo*	
BR1	0·15	0·26	2·48	0·012	0·036	15·70	12·43	2·40	nil
BR2	0·14	0·28	2·48	0·011	0·044	15·70	12·48	2·50	cracks
BR3	0·14	0·29	2·54	0·011	0·050	15·95	12·26	2·55	cracks
BR4	0·16	0·26	2·44	0·011	0·060	15·80	11·83	2·40	cracks
BR5	0·14	0·27	2·62	0·016	0·064	15·70	11·90	2·50	cracks

Table IX

Effects of phosphorus on cracking in 18/8 and 25/20 stainless steels[105]

Flux No.	Welds in 18/8 steel P, %	Hot cracks	Welds in 25/20 steel P, %	Hot cracks
1	0·020	None	0·024	None
2	0·032	None	0·041	None
3	0·039	None	0·047	Present
4	0·057	None	0·072	Present
5	0·061	None	0·114	Present
6	0·240	None	0·250	Present
7	0·300	None	0·300	Present
8	About 0·6	Present	About 0·6	Present

Toropov[130] shows that adding 0·049% P to 15 Cr–25 Ni–6 Mo weld metal (av. 0·76% C, 0·85% Si) produced cracks which could be seen with the unaided eye (Table VI).

In 25 Cr–20 Ni alloy, Medovar[105] found that cracks were present when the phosphorus content reached 0·047% (Table IX) but in 18 Cr–8 Ni steels no cracks were observed up to 0·30% P; at 0·6%, however, cracks were observed (Table IX). Adding 2·5% P to the 25 Cr–20 Ni steel prevented the formation of hot cracks. Both stainless steels were fully austenitic at room temperature, but Medovar does not state whether the 18 Cr–8 Ni was partially ferritic at high temperature. It is likely that this was so. Medovar[105] suggests that the greater sensitivity of 25 Cr–20 Ni welds (as opposed to 18/8) to hot cracking is due to the decrease in solidification temperature of the phosphorus eutectic, which contains a higher proportion of nickel than in the 18/8 type.

In Type 347 steel, phosphorus is also a potent crack promoter.[152] Increasing the phosphorus content from 0·028% to 0·044% increased the severity of cracking (segmented groove test) from 42·0% to 82·2% (Fig. 8).

Carbon and Silicon

The effects of carbon and silicon on weld cracking are somewhat inter-related and are therefore considered together. The effect of niobium on cracking is modified by the presence of carbon and silicon, but there are important differences that necessitate niobium being discussed separately.

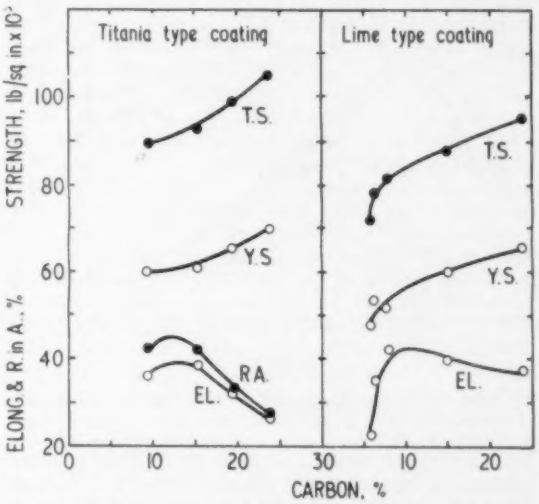
In studying Type 307 (19 Cr–9 Ni–Mn) and Type 308 (20 Cr–10 Ni) weld metal, Lee[34] concluded that cracking susceptibility depended on the combined effects of carbon and phosphorus. In presenting some of his results he assigned to each core wire a cracking index. This was an arbitrary number derived from the amount of extra chromium which had to be added to the core wire to prevent root bead cracking (Table X).

Table X

Effect of carbon on cracking in austenitic weld metal with similar Cr/Ni ratios[34]

Group Carbon, %	No. of Heats	Composition, % P	S	Cr	Ni	Cr/Ni Ratio	Cracking Index
0·05–0·07	16	0·01	0·011	20·7	9·8	2·12	65
0·08–0·09	29	0·019	0·011	20·5	9·7	2·12	67
0·10	16	0·020	0·013	20·7	9·8	2·12	69
0·11	19	0·021	0·013	20·9	9·6	2·13	75
0·12	42	0·020	0·011	20·8	9·9	2·09	98
0·13	28	0·021	0·012	20·7	9·8	2·10	90
0·14–0·15	21	0·023	0·013	20·6	9·7	2·12	136

In an investigation of the effect of alloying elements on the tensile properties of 25 Cr–20 Ni (Type 310) weld metal containing 0·42% Si, Campbell and Thomas[29] found that the addition of carbon from both lime-type and titania-type electrodes strengthened the weld metal and caused a corresponding decrease in ductility (Fig. 11). Optimum ductility was experienced at about 0·10–0·15% C. Linnert and Bloom[29] in discussing these results, compared them with similar results gained in their own laboratories and further



11—*Effect of carbon on mechanical properties of 25 Cr–20 Ni type weld metal*[29]

commented that intergranular fissures were noted at carbon contents below about 0·12% (Si content not given). In addition, they suggested that the low ductility values reported by Campbell and Thomas for weld metal containing less than 0·08% C indicated the presence of flaws.

Carpenter and Jessen[40] also investigated the properties of 25 Cr–20 Ni weld deposits, and as a result of their studies suggested that the preferred carbon range for weld deposits made with lime-coated electrodes was 0·12%–0·20%. Below this range lower tensile properties resulted; above it increased strength and lower elongations were experienced. Fissured weld metal was found at carbon contents of 0·12% or below, when Si exceeded 0·50%. At higher (0·18–0·20%) carbon levels 0·70% Si could be tolerated. They suggest that the harmful effects of silicon were due to the presence of silicate films around the austenite grains.

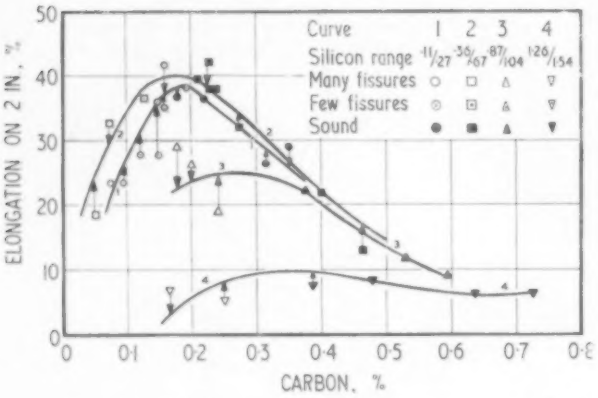
In a later paper Carpenter *et al*[63] suspected that carbon and silicon both caused a reduction in strength and ductility of weld metal deposited with 19 Cr–9 Ni electrodes (Table XI). Non-metallic films similar to

Table XI

Effects of carbon and silicon on the properties of fully austenitic 19 Cr–9 Ni weld metals in the as-welded condition[63]

Spec. No.	Yield Point, lb/sq.in.	Tensile Strength, lb/sq.in.	Elong., %	Red. in Area, %	Composition, % C	Si	Mn	S	P	Cr	Ni	Cr/Ni* Ratio	Fissures
E-121	54,000	70,750	13·0	27·8	0·09	0·35	2·19	0·015	0·015	17·74	13·99	1·27	Many
E-122	57,500	79,000	28·5	44·5	0·09	0·38	2·27	0·017	0·011	18·30	11·64	1·57	Few
E-654	38,250	58,640	23·0	44·7	0·07	0·75	1·92	0·010	0·012	18·03	12·64	1·43	Many
E-645	35,500	73,000	36·5	42·8	0·10	0·85	2·05	0·012	0·014	17·97	11·96	1·50	Many
E-646	45,500	91,000	43·0	57·0	0·17	0·41	2·04	0·012	0·012	18·43	8·92	2·07	None
E-647	51,500	82,250	53·0	62·2	0·18	0·48	2·02	0·016	0·012	18·74	9·12	2·05	None
E-648	49,000	80,000	47·0	64·4	0·17	0·62	2·14	0·015	0·012	18·63	9·48	1·97	Very few
E-649	48,000	79,000	41·5	39·8	0·18	0·70	2·06	0·012	0·016	18·21	9·40	1·94	Few

* Not included in original table



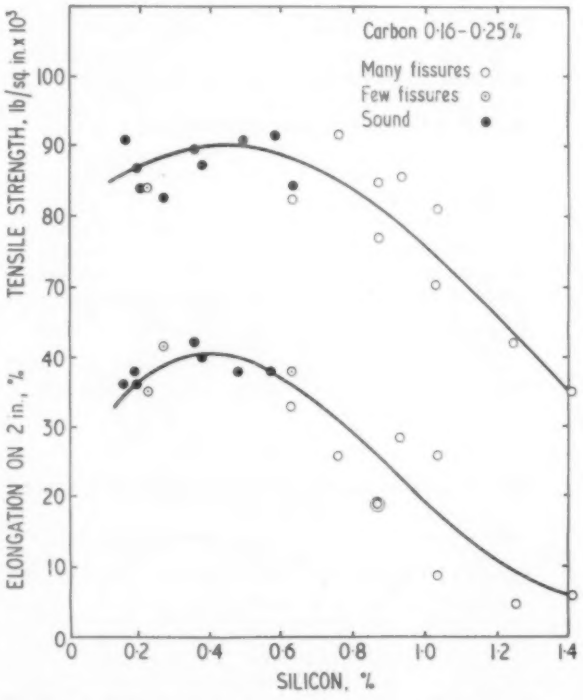
12—*Effect of carbon on ductility of 25 Cr–20 Ni weld metal*[43]

those described in the 25 Cr–20 Ni weld metal were believed to be responsible for cracking. In discussing this work, Linnert[87] thought that the Cr/Ni ratio was a more important factor than either carbon or silicon. He stated that the high Cr/Ni ratio of the last four weld metal compositions (E.646–E.649) indicated that they contained ferrite which accounted for their freedom from fissuring.

Campbell and Thomas[29] also observed that excessive Si contents in 25 Cr–20 Ni weld metal caused fissuring but, as previously mentioned, additionally noted that when present with phosphorus its effects were additive. They suggested that with 0·40% Si, the phosphorus content should not exceed 0·024%, whilst with 0·60% Si phosphorus should be limited to 0·016%. At the same phosphorus levels lime-type coatings were much less sensitive to silicon addition than were the titania-type coatings. In these experiments the carbon content was about 0·124%–0·146%.

Rozet *et al*[45] investigated the effect of carbon on the strength and ductility of 15 Cr–35 Ni welds, and showed that the level of carbon which gave optimum elongation values varied with the silicon level, and for Si contents below 0·6% was equal to about 0·20%. Below this figure fissuring was present and was the cause of low ductility values, whilst above 0·20% the weld remained sound, although ductility decreased. At 1·5% Si about 0·40% C was necessary to eliminate fissures (Figs. 12 and 13). A more precise relationship between C and Si is shown in Fig. 14. These authors believed that silicate slag films were responsible for the severe fissuring, but suggested that at the higher carbon levels the greater fluidity of the molten pool accounted for the disappearance of the slag films. In borderline specimens (fissure/no-fissures) where the presence of films was doubtful, they found that quench-annealing had no effect on the degree of fissuring, and thus inferred that these defects were caused by intergranular voids created during the solidification of the weld metal.

The remarks of these authors do not make it clear whether the silicate films were responsible for the voids which expanded into fissures on further cooling, or whether some other unexplained event took place. It

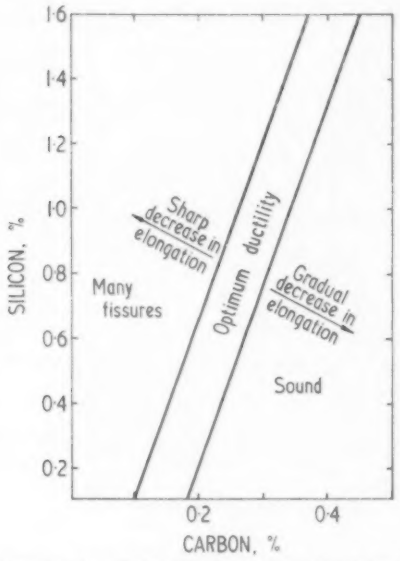


13—*Effect of silicon on mechanical properties of 15 Cr–35 Ni weld metal*[45]

is by no means certain also that brittle-type silicate films were not responsible for sub-solidus cracking, it being assumed that the defects were initiated below the solidus or 'effective solidus'.

Carpenter and Jessen[40] observed fissuring in Type 347 (19 Cr–9 Ni–1 Nb) manual arc weld deposits with a low carbon content (below 0·10%) despite the low silicon content (0·39%). Increasing the carbon to 0·18% eliminated all fissures and substantially improved the tensile properties, even with silicon contents as high as 0·70%.

Investigating the incidence of cracking using the Murex hot-cracking machine[113] in another Nb-bearing alloy, Rollason and Bystram[76] noted that a reduction



14—*Effect of carbon and silicon on optimum ductility of 15 Cr–35 Ni type weld metal*[45]

in cracking occurred at C/Si ratios exceeding 0·5. Carbon contents exceeding 0·09% had a pronounced effect. Omitting Nb in one alloy, they noted an increase in cracking, and therefore concluded that a high carbon content was not the only reason for improvement in hot cracking. The weld in question contained 0·11% C and 0·30% Si.

In welding an 18/8 steel (like Type 321), Medovar found that an increase in carbon content from 0·06–0·08% to 0·12–0·14% led to the appearance of hot cracks. The tendency for cracking increased when the weld contained Ti, Nb, or other energetic carbide formers. As the carbon content increased further, the tendency for hot cracking also increased, and then decreased; when the carbon content was 0·3–0·4% no cracks were found. The decrease in cracking was attributed to the refinement of the micro-structure as a result of the appearance of a carbide of the ledeburite type. The findings of Medovar appear to contradict those of Rollason and Bystram, since the latter authors found additions of carbon in a similar range to be beneficial. Perhaps this is not unexpected, because the 18/13/1 alloy was fully austenitic whereas the basic 18/8 type steel was partially ferritic. Adding carbon to this alloy would decrease the ferrite content, and might even make it fully austenitic.

Toropov[130] investigated the effect of carbon, at three levels of Si, on the cracking of fully austenitic welds containing normally 16 Cr–25 Ni–6 Mo. At low Si contents (0·15–0·21%) carbon up to 0·20% did not produce cracking; beyond this, hot cracking occurred but again disappeared when the carbon was above 1·5%. At medium and high Si contents (0·56–0·80% and 0·99–1·28%) an increase in carbon to 0·5–0·6%, increased hot cracking, whilst above this quantity there was no hot cracking (Table XII). Mechanical properties of uncracked weld metals containing medium and high Si contents were investigated, and it was found that the strength of weld metals containing high C and Si contents was higher but the ductility was lower.

Table XII

Effect of carbon on hot cracking in 15 Cr–25 Ni–6 Mo welds[130]

	Composition, %								
No.	*C*	*Si*	*Mn*	*S*	*P*	*Cr*	*Ni*	*Mo*	*Cracks*
Low silicon									
1	0·12	0·15	0·99	0·017	0·019	15·15	24·25	5·30	No
2	0·15	0·17	0·96	0·020	0·022	15·30	24·24	5·30	No
3	0·21	0·20	1·01	0·019	0·020	15·20	24·30	5·25	Yes
4	0·42	0·18	1·03	0·014	0·013	15·10	24·57	5·25	Yes
5	0·86	0·21	1·04	0·017	0·016	15·15	24·50	5·20	Yes
6	1·46	0·21	1·03	0·019	0·017	15·20	24·43	5·20	No
Medium silicon									
7	0·17	0·58	1·04	0·018	0·022	14·55	24·63	5·70	Yes
8	0·29	0·56	1·04	0·020	0·016	14·65	24·68	5·25	Yes
9	0·41	0·66	1·10	0·015	0·018	14·55	24·75	5·50	Yes
10	0·50	0·74	1·07	0·016	0·019	14·50	24·65	5·55	No
11	0·57	0·65	1·07	0·015	0·015	14·55	24·60	5·50	No
12	0·77	0·80	1·16	0·008	0·028	14·55	24·28	6·10	No
High silicon									
13	0·20	0·99	1·21	0·014	0·020	14·70	24·90	5·90	Yes
14	0·47	1·04	1·25	0·012	0·018	14·70	24·70	5·60	Yes
15	0·77	1·10	1·22	0·014	0·021	14·60	24·80	6·00	No
16	0·86	1·23	1·08	0·012	0·015	14·70	24·60	5·90	No
17	0·96	1·02	1·05	0·014	0·016	14·65	24·55	5·85	No
18	1·00	1·28	1·20	0·013	0·024	14·65	24·45	5·90	No

Metallographic studies revealed the presence of a grain-boundary precipitate in cracked high-C welds, but not in the uncracked samples. Uncracked welds with high carbon contents contained a continuous grain-boundary network. The intergranular substance was reminiscent of a carbide eutectic.

In investigating the welding of Type 347 steel, Hoerl and Moore[152] showed directly and statistically that increasing the carbon from 0·076 to 0·10% reduced cracking (segmented groove test). The niobium and tantalum content of the weld metal was about 0·62–0·88% (Fig. 8). They state that a somewhat surprising result was found in the study of the effect of Si on cracking. Although the range of variation was appreciable (0·32–0·73%) only a very slight increase in cracking with increasing Si content was recorded. They comment that their results are in contrast with the finding of other investigators, who had noted that Si was a major factor in promoting cracking. These findings however were only statistical, and direct experiments to confirm this statement were not carried out.

Niobium

Niobium is added to austenitic weld metal to stabilize it. Unfortunately, however, it tends to increase the susceptibility of fully austenitic weld metal to cracking. In large quantities it causes the formation of ferrite, which tends to prevent cracking.

The beneficial effects of adding Nb to 25 Cr–20 Ni weld metal containing excessive amounts of Si is explained by Campbell and Thomas[29] as being due to the appearance of ferrite in a previously fully austenitic structure.

Kihlgren and Lacy[31] investigated the effect of adding Nb to a range of Cr–Ni alloys, and they found that when the Nb/Si ratio exceeded a critical amount, cracks were not formed. These critical values were:

Alloy	*Critical Nb/Si Ratio*
80 Ni–20 Cr	3·8
80 Ni–14 Cr–6 Fe	4·5
60 Ni–15 Cr–25 Fe	7·0 (approx.)
35 Ni–15 Cr–50 Fe	8·0 (approx.)

Two probable reasons advanced by Kihlgren and Lacy for the deleterious effects of silicon were:

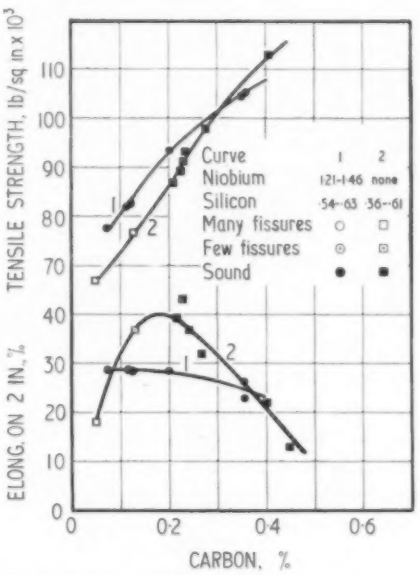
(i) The presence of an intergranular or interdendritic low-melting-point phase, and
(ii) The presence of a brittle continuous grain-boundary phase.

Whilst not having any evidence to explain the role of Nb, they theorized that a phase or intermetallic compound containing both Nb and Si was formed, the new phase being solid at the temperature at which cracking occurred in the unmodified weld metal. Niobium also possibly altered the equilibrium relationship, with consequent changes in the freezing range, or in the amount of and distribution of solidification shrinkage.

Carpenter and Jessen[40] also conducted experiments on the effects of adding Nb to 25 Cr–20 Ni weld metal, and they found that adding 2·11% in conjunction with 0·78% Si did not cause cracking. Metallographic examination revealed a fine grain size, and also the presence of a phase at the grain boundaries.

In investigating the properties of 15 Cr–35 Ni weld metal, Rozet, Campbell and Thomas[45] studied the effect of adding Nb (1·21–1·46%) in conjunction with Si (0·54–0·63%) and carbon (0·08–0·30%). These authors found that adding Nb to weld metal containing about 0·10% C eliminated fissures, but this addition was unnecessary if the carbon was increased to 0·20% (Fig. 15). However, since the ductility of the weld metal containing Nb was substantially lower than that of the alloys containing higher carbon contents, they concluded that niobium was an unnecessary addition.

Rollason and Bystram[76] in a study of the 18 Cr–13 Ni–1 Nb system, found that an improvement in cracking resistance occurred as the Nb/Si ratio increased (range studied up to Nb/Si=0·58). The addition of 1·0—1·5% Nb and of carbon in excess of 0·09% also had a beneficial effect. (It can be noted, however, from the table of compositions studied by these authors, that as the Nb/Si ratio increased so did the C/Si ratio. As previously mentioned, cracking is reduced as the C/Si ratio increases). These authors found that the microcracks were intercrystalline, and were formed mainly in weld metal reheated above 1250°C. They surmised that in the absence of niobium, $FeSi_2$ or FeSi with melting points about 1210°C. may form and lead to grain boundary weaknesses. The addition of more than 0·09% C reduced cracking.



15—*Effect of carbon and niobium on weld metal properties of 15 Cr–35 Ni steel*[29]

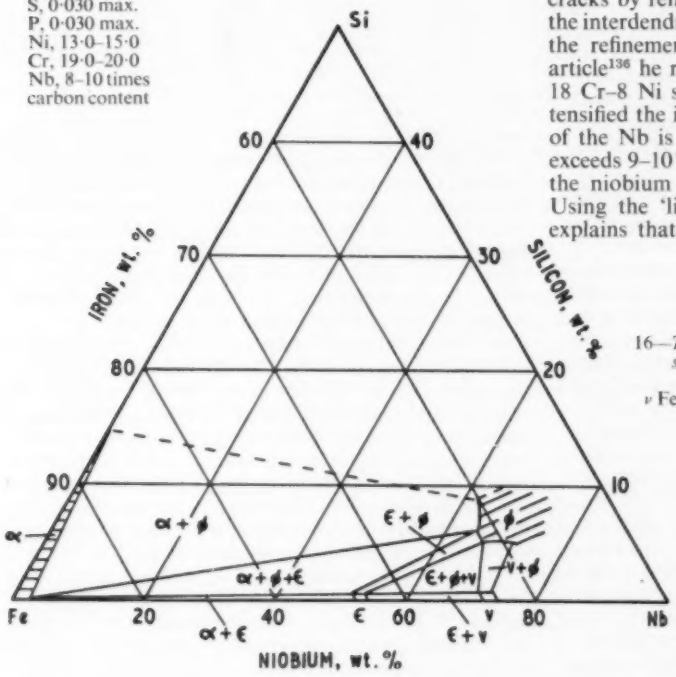
According to Bishop and Bailey,[84] who also investigated the welding characteristics of 19 Cr–13 Ni type weld metal, cracking occurred under the following conditions:

(*a*) The absence of delta ferrite from the structure
(*b*) The simultaneous presence of Si in amounts exceeding about 0·40% and of Nb in excess of about eight times the carbon content
(*c*) Reheating of the weld metal to high temperatures.

Metallographic evidence and subsequent X-ray

identification proved that a complex Fe–Nb–Si compound formed under conditions (*b*) and (*c*), but it led to cracking only in the absence of ferrite. A tentative specification for weld metal free from cracking was suggested as follows (wt.%):

C, 0·12–0·16
Mn, 0·7–1·4
Si, 0·30 max.
S, 0·030 max.
P, 0·030 max.
Ni, 13·0–15·0
Cr, 19·0–20·0
Nb, 8–10 times carbon content



16—*Tentative ternary diagram of the iron–niobium–silicon system. (Niobium content on double scale)*[84]
α Ferrite solid solution ε Fe_3Nb_2
ν Fe_xNb_y (probably $FeNb_2$) φ Silicide $Fe_4Nb_6Si_4$

In other papers,[105, 118] Medovar explains than an increased quantity of Ni–Si eutectic eliminates hot cracks by refinement of the structure and 'healing' of the interdendritic interstices. Delta-ferrite accompanies the refinement of the primary structure. In a later article[136] he reports that increasing the Nb content of 18 Cr–8 Ni steels from 0·15–0·20% to 0·6–0·8% intensified the incidence of cracking. The harmful effect of the Nb is made more severe if the nickel content exceeds 9–10%, or the carbon 0·08–0·10%. Increasing the niobium to over 1·0–2·0% eliminated cracking. Using the 'liquid healing' theory again, the author explains that at low Nb contents hot cracks cannot form because of the small amount of liquid which is present. Increasing the Nb content further causes cracks owing to the increase in the freezing range, and the fact that sufficient quantities of liquid films are developed. Still further increases reduce cracking due to 'eutectic healing'. This healing effect is accompanied by the appearance of ferrite in the structure, which is also beneficial because of its grain refining action. It is understood that the ability of ferrite to dissolve niobium and thus reduce segregation is not an important factor.

In a discussion[84] on their paper, Goldschmidt confirmed that excessive silicon in Nb-bearing material was an important factor in promoting cracking, and explained that it had been noted by X-ray examination that a Fe-Nb–silicide[68] of approximate formula $Fe_4Nb_5Si_3$ played a prominent role (Fig. 16). In the presence of carbon, the balance of elements decided whether Nb formed its carbide or its silicide.

Medovar[94] in a short review on the hot cracking of stainless steel suggested that in alloys containing over 15% Ni, silicon caused cracks, and that this was due possibly to the formation of a low-melting-point eutectic (1152°C.) with nickel. This author also refers to other Russian work where it had been found that Si counteracted the detrimental influence of Nb in 18/8 type steels. These alloys normally contain a small amount of ferrite and this fact may partially explain the apparent contradiction to Bishop and Bailey's findings. In fully austenitic alloys, Medovar believes that Si is detrimental unless in sufficient quantities to form ferrite.

A slightly different view of the reduction of cracking due to increased niobium content is taken by Lyubavskii and Toropov.[135] These authors found that cracks were present in 15 Cr–25 Ni–6 Mo (0·15–0·21% Si)

Table XIII

Influence of niobium on hot cracking in 15 Cr–25 Ni–6 Mo weld metal[135]

Test Pieces	*Composition, %*									*Presence of Cracks*
	C	*Si*	*Mn*	*S*	*P*	*Cr*	*Ni*	*Mo*	*Nb*	
AN1	0·12	0·15	1·08	0·010	0·019	15·35	24·60	5·38	0·15	Nil
AN2	0·13	0·17	1·10	0·008	0·018	15·30	24·45	5·42	0·27	cracks
AN3	0·14	0·18	1·05	0·009	0·018	15·30	24·40	5·40	0·65	cracks
AN4	0·16	0·20	1·10	0·011	0·019	15·36	24·48	5·40	1·06	cracks
AN5	0·16	0·21	1·11	0·012	0·018	15·25	24·35	5·45	1·68	cracks

Table XIV

Influence of niobium on hot cracking in 16 Cr–13 Ni–2 Mo weld metal[135]

Test Pieces	*Composition, %*									*Presence of Cracks*
	C	*Si*	*Mn*	*S*	*P*	*Cr*	*Ni*	*Mo*	*Nb*	
BN1	0·12	0·22	2·56	0·010	0·020	15·32	11·85	2·40	0·26	Nil
BN2	0·13	0·23	2·47	0·011	0·019	15·25	11·90	2·38	0·35	Nil
BN3	0·12	0·23	2·47	0·011	0·018	15·30	11·95	2·40	0·42	cracks
BN4	0·13	0·25	2·51	0·009	0·021	15·29	11·95	2·45	0·60	cracks
BN5	0·14	0·28	2·48	0·010	0·020	15·25	11·90	2·45	0·95	cracks
BN6	0·13	0·27	2·49	0·011	0·020	15·28	11·70	2·39	1·37	cracks
BN7	0·14	0·29	2·50	0·010	0·022	15·30	11·75	2·37	1·57	Nil
BN8	0·14	0·30	2·52	0·012	0·021	15·28	11·78	2·40	1·93	Nil

welds containing 0·27–1·68% Nb (Table XIII). But in type 16 Cr–13 Ni–2 Mo cracks were not found for niobium contents less than 0·35% nor were they formed when the amount exceeded 1·57%. Cracks were formed, however, in weld metal containing 0·42–1·37% Nb (Table XIV).

In the 16 Cr–13 Ni–2 Mo weld metal containing up to 0·35% Nb, the structure after welding was fully austenitic, with a small amount of carbides at the grain boundaries. In weld metal which cracked with a higher Nb content (0·95%) the structure was austenitic, with again a small amount of carbide, but in addition it contained a certain amount of intermetallic niobium compound, all mainly located at the grain boundaries. The weld with 1·37% Nb contained intermetallic niobium compounds and a small proportion of ferrite. The structure of welds containing over 1·5% Nb and without cracks was austenitic with ferrite regions and a small amount of carbides. From these results, and other heat-treatment studies, the authors concluded that the cause of cracking at average niobium concentrations was the formation of low-melting-point segregates (the authors in fact say "the reduction in melting point of the base material caused by segregation effects"). At the higher niobium contents cracking was prevented by the formation of ferrite, which was able to reduce segregation by reason of its greater powers of dissolving niobium.

In type 15 Cr–25 Ni–6 Mo weld metal, cracking was caused by segregation; that higher proportions of Nb did not prevent cracking is explained by the absence of ferrite and presumably (present authors comment) the amount (1·68%) introduced was insufficient to cause 'healing'.

Manganese

Manganese is generally considered to be an austenite former, but recent experiments suggest that it can act as a ferrite stabilizer when present in amounts exceeding 8%.[158]

The addition of Mn (0·93–2·93%) to 25 Cr–20 Ni (0·01% S) weld metal causes little or no change in the weld strength, and only slightly improves the ductility.[29] The minor effect of Mn on 25 Cr–20 Ni weld metal properties is also confirmed by Carpenter and Jessen.[40]

Rozet, Campbell, and Thomas[45] also found that manganese had an insignificant effect on 15 Cr–35 Ni weld metal properties (0·012% S). No fissures were observed.

Medovar and Latash[138] suggest that manganese increases the danger of cracking in welding 18 Cr–8 Ni steels. This rule, however, does not hold if, when 5–8% Mn is introduced, some ferrite phase is retained. Adding manganese to 25 Cr–20 Ni weld metal improves the resistance to cracking by combining with the sulphur. With this fact in mind Medovar and Latash suggest that part of the Ni content could be replaced by Mn. Thus they comment that instead of 20% Ni, not more than 15–16% Ni plus 6–8% Mn would be a desirable composition.

In the presence of copper, however, manganese is detrimental because of (1) the formation of low-melting-point ternary Ni–Mn–Cu sulphide, and (2) the reduction of the solubility of copper in austenite of increased manganese content, and the segregation of a low-melting-point Cu-rich phase to the columnar crystal boundaries. When copper is present in addition to sulphur, the manganese does not act as a neutralizer. In the absence of copper, however, manganese prevents the detrimental effect of sulphur (Table XV).

The benefits gained by adding manganese to fully austenitic weld metal were also realized by Kauhausen.[79] In 1951 he recommended the following weld metal composition for steam power boilers designed to operate at 1130°F. (610°C.):

C, %	0·115	Si, %	0·28
Mn, %	4·52	Cr, %	15·7
S, %	0·011	Ni, %	12·5
		Nb (Ta), %	0·90

This weld metal was claimed to be immune to hot cracking, but was susceptible to intergranular corrosion in the copper-sulphate–sulphuric-acid test after an exposure at 1200°F. (649°C.) for 2 hr.

Table XV

Effect of manganese, copper, and sulphur on hot cracking in welding stable austenitic steels[138]

No.	*Additional Alloy in Weld*	*Steel Quality*	*Presence of Copper in Steel*	*Presence of Copper in Weld*	*Cracks in Weld*
1	None	EI 533	Yes	Yes	No
2		EI 628	No	No	No
3	0·1% S	EI 533	Yes	Yes	Yes
4		EI 628	No	No	Yes
5	5% Mn	EI 533	Yes	Yes	Yes
6		EI 628	No	No	No
7	0·1%S+5%Mn	EI 533	Yes	Yes	Yes
8		EI 628	No	No	No
9	5%Mn+2%Cu	EI 528	Yes	Yes	Yes
10	2%Cu	EI 628	Yes	Yes	No
11	0·1%S+5%Mn +2%Cu	EI 628	Yes	Yes	Yes

In a later article (1955) Kauhausen and Vogels[115] indicated that this weld metal composition was not used to any great extent in the 1130°F. steam boiler application.

Molybdenum

Thomas[28] relates the beneficial effect of molybdenum added to 19 Cr–9 Ni weld metal to the presence of ferrite found in the structure, as does Tremlett[100] in investigating the cracking of 18 Cr–8 Ni–Mo type steel, though even the fully austenitic Type 316 (18 Cr–13 Ni–Mo) is highly resistant to cracking.[28] Molybdenum also reduced the cracking susceptibility of austenitic 20 Cr–10 Ni steel weld metal.[22] Generally, Mo-bearing austenitic weld metals are less susceptible to hot-short cracking than balanced compositions of weld metal containing no molybdenum. However, a combination of Mo and Nb seems to increase the tendency to hot tearing and cracking.[53]

Lyubavskii and Toropov[135] suggest that molybdenum, when added to 16 Cr–13 Ni and 15 Cr–25 Ni weld metals should not cause cracking since this element is very soluble in Fe, Cr, and Ni, and has little effect on their melting points. Molybdenum to 3·10% in 15 Cr–25 Ni welds however caused the formation of sigma. In 16 Cr–13 Ni welds these authors found that 9·5% Mo did not cause sigma.

Copper

In Nb-bearing 18 Cr–8 Ni–2 Cu weld deposits microcracks are usually absent, but should the first pass of a multipass weld be stress-relieved at about 850°C. then it is possible that cracks may form in the first pass, if further welding be carried out.[73] The reason for this is the low ductility of the weld metal caused by intergranular precipitates of the ϵ-phase (Fe–Cu system) resulting from long exposures at 850°C. The weld thermal cycle of the first run is not sufficiently prolonged to cause any ill effects.

Adding 2–14% Cu to Type 316 (18 Cr–12 Ni–3 Mo) weld deposits, Williams *et al*[39] found that the cracking increased markedly. Weld/base-metal cracks were also noted. The base metal composition was 16 Cr–25 Ni–6 Mo.

Vanadium

Lyubavskii and Toropov[135] find that adding 0·82–1·76% V to 15 Cr–25 Ni–6 Mo weld metal causes cracks (Table XVI) but additions up to 1·4% V to weld metal containing 15 Cr–12 Ni–2 Mo do not. Medovar[136] cannot agree with the suggestion that vanadium causes hot cracks, but no test results are given.

Zirconium

Medovar[136] finds that adding 0·25–0·75% Zr to nickel causes hot cracks during welding, and that Zr also behaves similarly in the welding of high-nickel austenitic steels, but again no tests results are quoted.

Hydrogen

Hydrogen has been suggested as a cause of cracking in 18 Cr–8 Ni type welds.[55] In many experiments however, Medovar[94] could not detect any appreciable influence of hydrogen on hot cracking. He suggests that hydrogen possibly intensifies cracking, caused by other elements, by its manner of diffusing into the cracks and increasing the internal pressure as molecular hydrogen is formed.

Oxygen

In 18 Cr–8 Ni steels, cracks can be caused by the presence of oxygen.[94] Medovar interprets the harmful effect of oxygen, not with a change of composition of the grain-boundary layers nor with non-metallic inclusions, but with the fact that the weld is austenized by the oxidation of ferrite-forming additions. It has also been thought[47, 80] that ferrite formers reduce cracking not because delta ferrite is formed, but because these elements are powerful deoxidizers. It was suggested[47, 80] that deoxidizing elements, such as titanium, might ameliorate the effect of oxygen. Some experiments by Medovar however show that Ca, Mg and Zr, which are energetic deoxidizers but not ferrite formers, are not capable of suppressing cracks in 18 Cr–8 Ni and 25 Cr–20 Ni weld metals. Furthermore, although V is a less powerful deoxidizer than Si, hot cracks in a weld metal containing 18% Cr, 11–12% Ni, 0·10% C, and 0·5% Mn can be eliminated by adding only 1·0–1·4% V, whereas 1·8–2·0% Si is necessary to have the same effect.

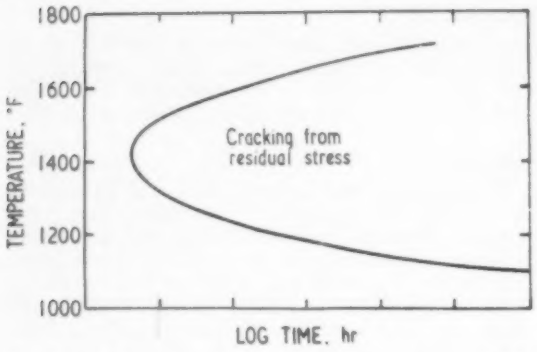
Nitrogen

Since nitrogen is an austenizer and a potent strengthener, it might be expected that it would be harmful when added to stainless steels. This however, does not appear to be the case universally. The nitrogen content of 18 Cr–8 Ni stainless steel is usually about 0·05–0·06% and in such amounts it does not exert a harmful influence on the cracking properties of austenitic-ferritic welds, except when Nb is also present; this is in spite of the fact that nitrogen tends to refine the structure.[94] When niobium is present, nitrogen in amounts less than about 0·05% greatly intensifies the cracking.[94] If it is absent, the effect of increasing the nitrogen content from 0·02% to about 0·06% is

Table XVI

Effect of vanadium on 15 Cr–12 Ni–6 Mo weld metal[135]

Specimens	*Composition, %*									*Presence of Cracks*
	C	*Si*	*Mn*	*S*	*P*	*Cr*	*Ni*	*Mo*	*V*	
AV1	0·12	0·15	1·12	0·012	0·015	15·51	24·95	5·60	0·32	Nil
AV2	0·12	0·14	1·13	0·010	0·012	15·60	24·90	5·66	0·51	Nil
AV3	0·13	0·16	1·12	0·014	0·011	15·57	24·87	5·63	0·82	cracks
AV4	0·12	0·15	1·15	0·013	0·013	15·52	24·89	5·67	1·23	cracks
AV5	0·13	0·16	1·14	0·012	0·012	15·55	24·90	5·65	1·76	cracks



17—*Hypothetical curve showing time at temperature needed for residual stresses to cause cracking of Type 347 weldments during post-weld heat-treatment*[124]

beneficial, and may even eliminate cracking if the austenite/ferrite structure is changed from an orientated one into a random one by the grain refining action of the nitrogen. An 18/8 weld metal which was austenitic because it contained 0·10% nitrogen contained cracks.[94]

Post Weld Heat Treatment

Heat treatment after welding is usually carried out for three good reasons: (i) To relieve residual stress;[114,155] (ii) to modify the weld microstructure—this usually means a solution treatment at 1900°–2000°F. (1037°–1093°C.) to remove delta ferrite from the weld metal and to coalesce the pools of this phase which remain;[124] (iii) to improve the corrosion resistance. In an interpretive report on the stress relieving of stainless steels Huseby[155] divides post-weld treatments into three broad temperature ranges. These are (i) 800°–1600°F. (427°–971°C.), which was indicated to be the sensitizing range; (ii) 1350°–1750°F. (732°–954°C.)—the stabilizing range; and (iii) 1750°–2050°F. (954°–1121°C.), which was described as the annealing range.

The employment of temperatures higher than 2000°F. (1093°C.) are not recommended because of scaling, distortion, and solution of Nb and Ta carbides. Subsequent exposure of Type 347 steel to temperatures of 1200°F. (649°C.), such as in steam piping applications, would precipitate the carbon as chromium carbides.

There is no conclusive evidence that cracks are originated by stress-relieving treatments, or during service, but experience and experiments have shown that welds may contain numerous cracks after stress relieving or after service conditions have been sustained.

Curran and Rankin[124] comment that cracking of Type 347 austenitic welds during post-weld heat treatment is accentuated by slow rates of heating, and that these observations indicate that cracking is due to the residual welding stresses rather than to stresses caused by uneven heating. They suggest that there is a possibility of a stress rupture failure occurring during a stress-relieving treatment. Whether cracking occurs or not depends on the relative rates of decrease of residual stress[60] and of rupture strength during post-weld heat treatment. Using this argument they hypothesise that a plot of cracking time versus temperature would be of the general shape shown in Fig. 17.

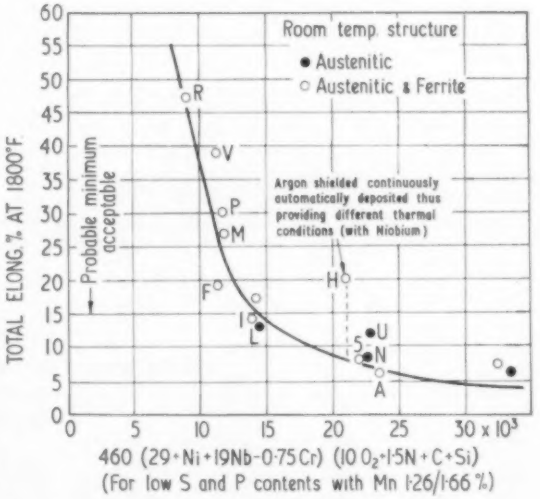
Curran and Rankin subjected several highly restrained Type 347 weld metals to the following heat treatment:

(i) Heat to 1100°F. (593°C.) at 300°F./hr (170°C./hr)
(ii) Hold at 1100°F. for 2 hr to permit temperature equalisation and relief of the higher residual stresses
(iii) Heat from 1100° to 1925°F. (1050°C.) at 600°F/hr (330°C.)/hr
(iv) Hold 2 hr for complete stress relief and ferrite solution
(v) Air cool.

All the test weldments subjected to this heat-treatment cycle were found to be free from cracking when inspected after the air cool.

Effect of Temperature, Stress and Time on Weld Metal Properties and the Possible Relationship to Weld Cracking

Heuschkel[125] reports the temperature-strength and temperature-ductility weld metal properties of several Cr–Ni austenitic compositions (Types 308, 309, 310, 312, 316, 321, 347 and 349). The welds he tested were made under conditions of minimum restraint to avoid 'after weld' cracking. Short-time tensile tests up to 2200°F. (1204°C.) were conducted, and yield and tensile strength, reduction in area, uniform elongation, and total elongation measurements were recorded. In particular, the elongation at 1800°F. (the high-temperature brittleness) was noted and shown to be inversely proportional to the matrix strengthening and the grain-boundary weakening components of the composition under investigation (Fig. 18). The high-temperature brittleness effect was considered to be important, since it was known that cracking had been experienced in welds which exhibited low ductility at 1800°F. Even though cracks did not necessarily form



18—*Relationship between product of strengthening and segregating components of compositions and 1800°F. elongation for AISI 300 series as-deposited weld metal*[125]

immediately after welding, if the material exhibited high-temperature brittleness cracks might occur after heat treatment or during service under stress at temperature as low at 1050°F. (566°C.).

Heuschkel's results are summarized under the following headings:

Group I—Austenitic Structures

At 1800°F. (980°C.) no high ductility welds were secured with existing commercial electrodes, when the as-deposited structure was austenitic at room temperature. As the initial structure of the plain Cr–Ni steel welds approached the fully austenitic condition, the total elongation approached a minimum (Fig. 19), a reverse of the condition found for strength (Fig. 20). Heuschkel postulates that these materials are crack sensitive, not because they are austenitic at room temperature, but rather because when they are austenitic at room temperature, the grain matrix is relatively strong at such temperatures as 1800°F. (982°C.). This allows a high concentration of stress on the contaminated grain boundaries.

Group II—9% Ferrite Content

The 20 Cr–10 Ni and 25 Cr–14 Ni welds contained 9% ferrite at room temperature. They were less strong and more ductile than the austenitic welds at the higher temperatures (Figs. 19 and 20).

Group III—27% Ferrite Content

The 27 Cr–9 Ni weld was the hardest and least ductile at and below 900°F. (482°C.), and was of average ductility at 1200°F. (640°C.). It was the least strong and most ductile at and above 1800°F. (982°C.).

Group IV—Titanium and Molybdenum Additions

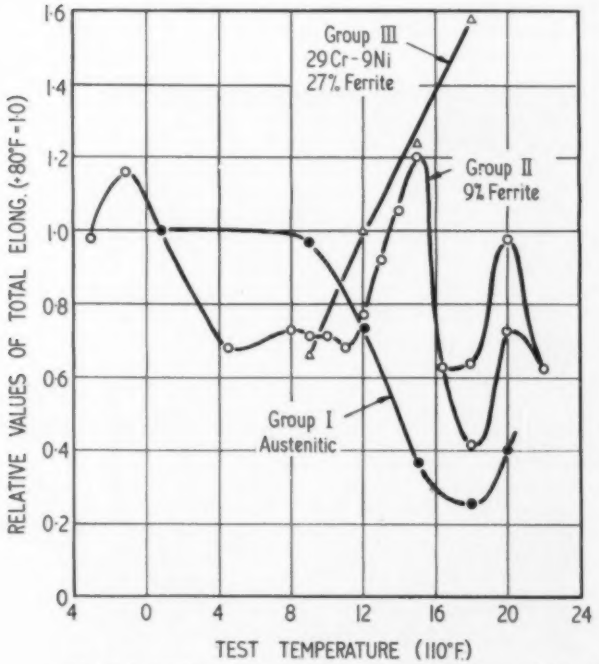
For a given high-temperature strength, these additions generally result in improved high-temperature ductility.

Group V—Niobium Additions

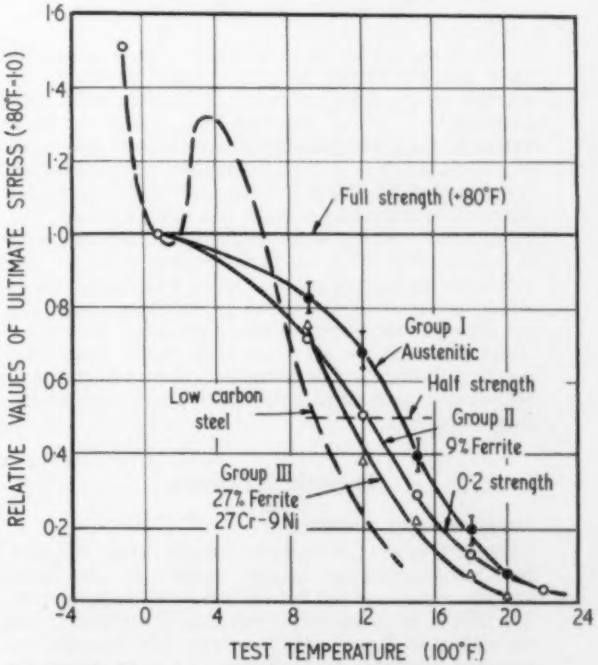
Heuschkel suggests that high-temperature brittleness explains why service cracking is often met. Best results were secured with inert-gas shielding and combined Mo and W additions when lower niobium, oxygen and silicon contents were used.

In general, in the Cr–Ni steel weld metal, Ni and Nb provide major sources of high-temperature matrix strength, with Cr additions tending to lower that strength. Better high-temperature ductility is secured for a given strength level when alloying additions of Mn, Ti, Mo, or W are made. A similar effect is secured with Nb but this is such an effective strengthener that, for rapidly quenched conditions, welds containing this element tend to become overstrong and insufficiently ductile at temperatures greater than 1500°F. (815°C.). This is particularly true for electrodes containing a minimum Nb content of eight times the carbon content, with a 1% maximum.

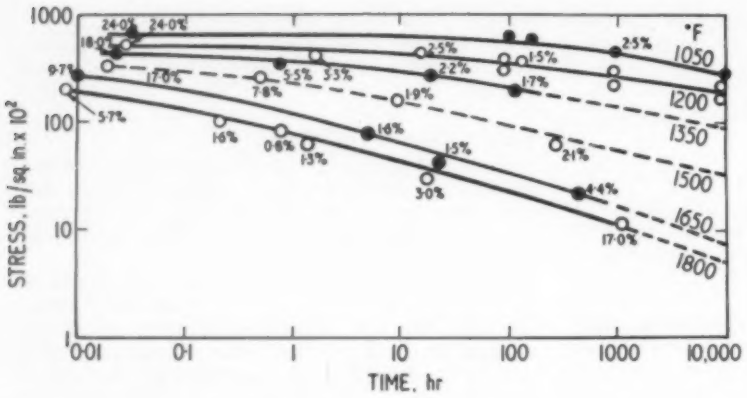
In a later paper Heuschkel[133] states that Type 347 weld metal having the stress rupture properties shown in Fig. 21 might easily crack if given a solution treatment of 1950°F. (1066°C.) or higher. This is because, on heating through the 1800°F. range, the fracture strain would be exceeded. He also comments that high-ferrite weld metal should be avoided because of the tendency to form sigma or carbides, or both. The best properties would be secured with a low ferrite (2–6%) weld metal having a minimum short-time tensile



19—*Relationships between temperature, elongation, and initial ferrite content for Cr–Ni welds (without niobium)*[125]



20—*Relationships between temperature, ultimate stress, and initial ferrite contents for Cr–Ni welds (without niobium)*[125]



21—*Stress-rupture data for Type 347 weld metal*[133]

elongation of about 15% at 1800°F. (See also ref. 128 for discussion on "Dependence of steel properties on lattice structure",)

Sigma Formation

There is no apparent relationship between cracking susceptibility and the amount, form, and distribution of sigma in weld deposits. It is known that sigma can adversely affect the room-temperature mechanical properties of welds, but whether or not the formation of sigma can cause cracking at elevated or service temperatures is not clear.[65,78,104,117,123,124,148] Fabricators do not experience troubles with low ferrite (1–6%) weld deposits.[124]

BASE METAL CRACKING

The literature up to 1950 makes little reference to base metal cracking. In recent years, however, the number of incidents of cracking has increased, particularly in Type 347 material. This is due to a large extent to the more widespread use of Cr–Ni austenitic steels, particularly in high-temperature applications.[72]

According to Poole,[98] who carried out a survey among 23 fabricators of Type 347 steel in the U.S.A., base-metal cracking presents a serious problem, since it occurs most often in sections thicker than $\frac{3}{4}$ in., which are difficult to repair. Unlike weld-metal cracks which are generally assumed to occur at some time during the weld thermal cycle, base-metal cracks are often observed for the first time some time after welding. These are due either to stress relieving or to high-temperature service conditions, or to a combination of the two.

Incidence of Cracking

Immediately after Welding

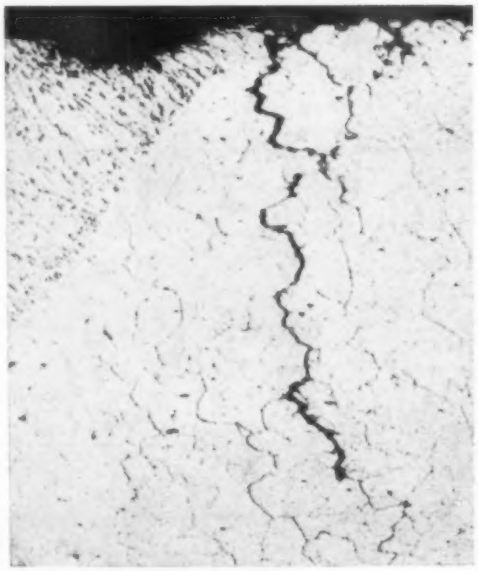
Poole[98] in his survey notes that in Type 347 steel base-metal cracking usually occurs in the heat-affected zone, and for the most part is intergranular. The cracks are usually, but not always, parallel to the line of fusion. Cracking in coarse grained forgings was reported to occur at distances of $\frac{1}{8}$–$\frac{3}{4}$ in. from the fusion line, and to propagate in a direction normal to the surface. No mention is made of fabricators finding continuous weld/base-metal cracks.

Puzak, Apblett and Pellini,[126] in investigating the welding properties of castings and heavy forgings in Type 347 steel, noted that base metal, weld/base-metal, and fusion-line cracks all occurred.

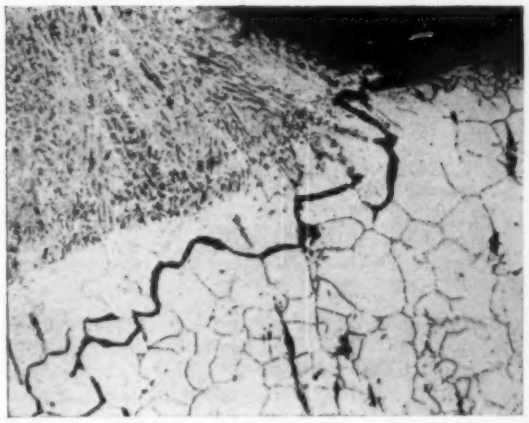
Medovar[147] reports several examples of weld/base-metal cracks in 23 Cr–27 Ni–Mo–Ti material when Cu (3·0%) or S are added to the weld metal, and suggests that some of these are due to Cu-rich weld metal penetrating between the grains of the base metal. Weld metal containing 15% Cu did not cause cracks in the base metal because of the 'liquid healing' that occurred.

After stress relief

Extensive cracking of base metal regions may be developed during, or subsequent to stress relieving or full heat treatments of welded assemblies.[126] Poole[98]



22—*Weld heat-affected zone cracking* ×150
(Courtesy International Nickel Company)



23—*Heat-affected zone cracking in 18|13|1Nb material. (Austenitic steel base-metal cracking test)*

states that cracking during stress relief has come with the advent of heavier sections and occurs in both base and weld metals which were, before stress relief, apparently sound by the best means of non-destructive testing. The nature and occurrence of the cracks is largely similar to cracks seen in the as-welded condition. One instance of transgranular cracking perpendicular to fusion line was noted, and although the reasons for this cracking were not established, it was thought that ferrite stringers were responsible.

After high-temperature service

Soldan and Mayne[144] report that, in heavy-wall Type 347 stainless steel piping, the percentage of failures in service is very small. In an examination of these failures, after initial fabrication and following service, they found that most of the cracking lay in the outer third of the pipe wall. The cracking usually appeared to have started at the outer surface and progressed normal to the pipe wall (Fig. 22). Repairs to these cracks were difficult to make because of progressive cracking further back.

That these materials are liable to failure in high-temperature service conditions is confirmed by other workers.[149–151, 158]

Evaluation of base-metal cracking tendencies

The methods which have been used for determining the suitability of materials for welding are as follows:

Table XVIII

Base metals, electrodes, and heat treatments used in welded pipe joints subjected to elevated-temperature testing[143]

Test No.	*Base Metal*	*Heat Treatment*	*Welding Electrodes*	*Postweld Heat Treatment*
8 and 13	347	As-received	347	As-welded
	347	As-received		
25	347	1900°F., WQ	347	As-welded
	347	1900°F., WQ		
18 and 26	347	2100°F., WQ	347	As-welded
	347	2100°F., WQ		
	347	2300°F., WQ	347	As-welded
	347	2300°F., WQ		
21 and 22	347	2375°F., WQ	347	As-welded
	347	2375°F., WQ		
10	347	As-received	347	1900°F.–2 hr–air cool
	347	As-received		
	347	1900°F., WQ	347	1900°F.–2 hr–air cool
	347	1900°F., WQ		
27	347	2100°F., WQ	347	1900°F.–2 hr–air cool
	347	2100°F., WQ		
	347	2300°F., WQ	347	1900°F.–2 hr–air cool
	347	2300°F., WQ		
	347	2375°F., WQ	347	1900°F.–2 hr–air cool
	347	2375°F., WQ		
9	316	As-received	316	As-welded
	316	As-received		
12	316	1900°F., WQ	316	As-welded
	316	1900°F., WQ		
14	316	2100°F., WQ	316	As-welded
	316	2100°F., WQ		
16	316	2300°F., WQ	316	As-welded
	316	2300°F., WQ		
19	316	2375°F., WQ	316	As-welded
	316	2375°F., WQ		
11	316	As-received	316	1900°F.–2 hr–air cool
	316	As-received		
	316	1900°F., WQ	316	1900°F.–2 hr–air cool
	316	1900°F., WQ		
15	316	2100°F., WQ	316	1900°F.–2 hr–air cool
	316	2100°F., WQ		
17	316	2300°F., WQ	316	1900°F.–2 hr–air cool
	316	2300°F., WQ		
20	316	2375°F., WQ	316	1900°F.–2 hr–air cool
	316	2375°F., WQ		
1	347	As-received	347	As-welded
	347	2100°F., WQ		
2 and 3	347	As-received	347	As-welded
	347	2300°F., WQ		
4	347	As-received	347	As-welded
	316	As-received		
5	347	As-received	347	As-welded
	316	2300°F., WQ		
6	347	2300°F., WQ	347	As-welded
	316	As-received		
7	316	As-received	316	As-welded
	316	2100°F., WQ		
24 and 28	347	1900°F., WQ	16–8–2	As-welded
	347	1900°F., WQ		
23	316	1900°F., WQ	16–8–2	As-welded
	316	1900°F., WQ		

Table XVII

Chemical analyses of materials[142] (see Tables XVIII–XIX)

Material Analysed	*Composition, %*						
	C	*Si*	*Mn*	*Cr*	*Ni*	*Mo*	*Nb*
Type 347 base metal	0·060	1·73	1·80	18·61	11·07	—	0·65
Type 316 base metal	0·080	0·15	1·64	17·64	10·68	2·30	—
Type 347 weld deposit (from Test 8)	0·059	0·43	1·75	20·41	12·08	—	0·78
Type 347 weld deposit (from Test 13)	0·073	0·44	2·00	20·33	10·72	—	0·73
Type 316 weld deposit (from Test 9)	0·073	0·53	1·59	19·54	14·42	2·55	—
Type 316 weld deposit (from Test 11)	0·064	0·33	1·72	19·71	15·70	2·75	—
16–8–2 weld deposit	0·086	0·45	1·65	15·87	8·37	1·42	—

Base-metal cracking tests: During welding

Apblett and Pellini[103] consider that base-metal cracking is associated with the presence of low-melting segregate films. To confirm this hypothesis they made melt runs on coarse grained samples of cast 18 Cr–8 Ni and 18 Cr–9 Ni–Nb material containing about 0·10% sulphur, and found that cracks were developed. Further experiments showed that fine grained material from the same casting and course grained material which was hot forged were not prone to base-metal cracking.

In a later investigation, Puzak, Apblett and Pellini[126] carried out a series of auto cracking tests to evaluate base-metal cracking properties of Types 304 and 347 material. These tests consisted simply of melt runs or electrode-deposited welds on a 3×6×1 in. section of the base material. The results of their studies are shown in Table XX. It is noted that standard 304 (0·09% C) and low carbon (0·03%) 304 material showed no tendency to develop base-metal cracks; this is ascribed to the combination of fine grain and a lack of fusible grain-boundary constituents. Extensive base-metal cracking was developed in the standard 347 material possessing large columnar grains, but no base-metal cracks were observed in the fine grained material In the low carbon (0·03%), low niobium (Nb+Ta= 0·50%) 347 material twice as many cracks were observed as in the standard 347 with the columnar grain structure. Forging the material to redistribute the fusible segregates decreased the length of the individual base-metal cracks but increased their number, as would be expected because of the great number of grain boundaries available for fusion.

The behaviour of the low carbon and high niobium (Nb+Ta=0·98%) 347 material was similar to that of the low carbon, low niobium 347 steel.

After heat-treatment and simulated high-temperature service

Emerson and Jackson[142] studied the plastic ductility of welded joints in Type 316 and Type 347 material, which had been previously heat treated, under simulated high-temperature conditions (1200°F., 649°C.). It was the authors intention to determine if failure could be made to occur in the base metal of stainless steel piping under laboratory conditions similar to those in actual service. The conditions of base materials, electrodes, and the heat treatments used are given in Tables XVII and XVIII and a summary of their results is presented in Table XIX. Of the many comments made and conclusions drawn by these authors, the following are most important:

(*a*) Weld heat-affected-zone failure in Type 347 occurs at the outside surface and progresses inwards
(*b*) At room temperature, Type 316 was more 'ductile' than Type 347 after heating each to 2300°F. (1260°C.), and thus it may be considered that the latter material is more susceptible to structural damage when subjected to high-temperature service conditions.

Borland and Younger,[162] using a simple cracking test, have shown that small hot cracks occur in Types 347 and 316 Nb steels under high restraint. Crack propagation during stress relief at 750°C. gives failures (Fig. 23) which appear to be similar to those in service (Type 347 only). Small hot tears seen after welding Type 316 material did not propagate under similar conditions of stress.

Hot ductility measurements

Since cracking is thought to occur at high temperatures, it is not unreasonable to suppose that the high-temperature properties[13,122] of materials are very important. Tests which evaluate the hot strength or ductility of metals are likely to correlate with the welding behaviour. It was probably this viewpoint which led Nippes and his co-workers[120,140] to design and construct apparatus with which to study the effect of rapid time/temperature cycles, associated with arc welding, on the hot ductility of austenitic stainless steels.

Preliminary experiments[120] by these workers were conducted on standard Type 316 Nb and 347 stainless steels, and the following conclusions were reached:

(i) The hot ductility of Type 347 steel measured on heating was far superior to that of Type 316 Nb at testing temperatures between 1400°F. (860°C.) and 2300°F. (1260°C.).
(ii) Prior thermal heating of the specimen had a marked effect: short exposure at a temperature of 2400°F. (1320°C.) caused a significant reduction in hot ductility for both grades at temperatures between 1400°F. (760°C.) and2 300°F. 1260°C.).

Table XX

Results of auto-cracking tests in types 304 and 347 steels[126]

Material	*Plate Grain Size*	*Cracking* (a) *due to Fusion Weld*		*Cracking* (a) *due to 4% Ferrite Weld*		*Cracking* (a) *due to 8% Ferrite Weld*	
		Base	*Weld*	*Base*	*Weld*	*Base*	*Weld*
Std. 304	Fine	None	None	None	None	None	None
Low C 304	Fine	None	None	None	None	None	None
Std. 347	LCT	18(*b*)	10(*b*)	19(*c*)	1(*c*)	16(*c*)	4(*c*)
	Fine	None	None	—	—	—	—
Low C 347 (low Nb)	LCT	39(*b*)	25(*b*)*	—	—	45(*c*)	13(*c*)
	Fine	48(*c*)	38(*b*)*	41(*c*)	26(*c*)	45(*c*)	11(*c*)
Low C 347 (high Nb)	LCT	52(*b*)	51(*b*)	—	—	70(*c*)	16(*c*)
	Fine	71(*c*)	69(*b*)*	37(*c*)	25(*c*)	47(*c*)	6(*c*)

* Also severe longitudinal weld cracking.

Notes: (*a*) Notch extension cracks were not visible to naked eye even on polished specimens; (*b*) Relatively large cracks visible as welded at ×30 magnification; (*c*) Hairline cracks visible only at ×125 magnification.

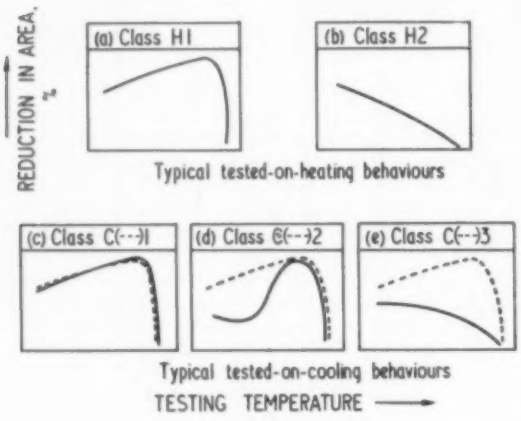
Table XIX

Summary of results of high-temperature tests made on welded heavy-wall pipe joints of AISI types 347 and 316 stainless Steel[112]

Test No.	Failure Location			Material	Heat Treatment before Testing: Base Metal Heat Treatment					Electrode and Postweld Heat Treatment		Remarks
	W*	HAZ†	BM‡		AR‖	1900	2100	2300	2375	AW§	1900	
8	x	—	—	347	347	—	—	—	—	347	—	Fuse blew out at 1200°F. allowing specimen to cool to room temperature. Centreline weld failure.
13	x	—	—	347	347	—	—	—	—	347	—	Centreline weld failure as Test 8 above but no heating failure.
25	x	—	—	347	—	347	—	—	—	347	—	On last (eleventh) heating specimen failed upon cooling to room temperature with bolts tightened
18	—	x	—	347	—	—	347	—	—	347	—	Specimen failed prematurely in HAZ of joint 4 in. from test weld.
26	x	—	—	347	—	—	347	—	—	347	—	Heater burned out and specimen cooled to 520°F. after 40 min at 1200°F.
21	—	x	x	347	—	—	—	—	347	347	—	Cracks present in base metal after heat treatment and before welding and testing. Failed in HAZ of joint 4 in. from test weld.
10	—	—	—	—	347	—	—	—	—	—	347	No failure.
27	—	—	—	—	—	—	347	—	—	—	347	No failure.
22	—	—	—	347	—	—	—	—	347	347	—	No test—cracks present in base metal after heat treatment and before welding. No postweld heat treatment
9	—	—	—	—	316	—	—	—	—	316	—	No external evidence of failure—sectioning revealed several weld metal microfissures believed to be present before testing.
12	—	—	—	—	—	316	—	—	—	316	—	No failure.
14	—	—	—	—	—	—	316	—	—	316	—	No failure.
16	—	x	x	316	—	—	—	316	—	316	—	Cracks at ID of 316 base metal as well as in HAZ.
19	—	x	x	316	—	—	—	—	316	316	—	Rupture cracks at HAZ as well as numerous others throughout base metal.
11	—	—	—	—	316	—	—	—	—	—	316	No failure.
15	—	—	—	—	—	—	316	—	—	—	316	No failure
17	—	—	—	316	—	—	—	316	—	—	316	No test: cracks found in HAZ before testing.
20	—	x	—	316	—	—	—	—	316	—	316	One crack about $\frac{1}{8}$ in. deep.
1	—	—	x	347 2100°F.	347	—	347	—	—	347	—	Failure in reduced section base metal at sharp-cornered fillet.
2	—	x	—	347 2300°F.	347	—	—	347	—	347	—	No remarks.
3	—	x	x	347 2300°F.	347	—	—	347	—	347	—	Rupture cracks starting at OD base metal at location of thermocouple tack welds.
4	—	x	—	347	347–316	—	—	—	—	347	—	Failure between butt weld and thermocouple tack weld.
5	—	x	x	316	347	—	—	316	—	347	—	Several rupture cracks starting at ID of 316 base metal.
6	—	x	x	347	316	—	—	347	—	347	—	Several rupture cracks starting at OD of 347 base metal at location of thermocouple tack welds.
7	—	—	—	—	316	—	316	—	—	316	—	No failure.
24	FL¶	—	—	—	—	347	—	—	—	16–8–2	—	Fusion line failure.
28	FL¶	—	—	—	—	347	—	—	—	16–8–2	—	Fusion line failure.
28	—	—	—	—	—	316	—	—	—	16–8–2	—	No failure.

*W = weld; †HAZ = heat-affected zone; ‡BM = base metal; ‖AR = as-received; §AW = as-welded; ¶FL = fusion line

In further work using the same apparatus, Nippes and his collaborators[140] studied the high-temperature properties of 34 different materials. The temperature/ductility curves exhibited characteristic trends (Fig. 24) and thus they were able to classify them and draw some general conclusions.



24—*Ductility curves*[140]

Class H1

Most materials on heating exhibited a gradual increase in ductility as the testing temperature was raised to a maximum (usually occurring in range 2200°–2400°F. (1200–1320°C.) depending on the alloy) followed by a marked loss in ductility over a narrow range of testing temperature 50°–150°F. (25°–75°C.) below the melting point of the alloy.

Class H2

A few materials tested on heating exhibited a gradual continuous decrease in ductility over a wide range of testing temperatures.

Class C1

The tested-on-cooling behaviour was similar to class H1.

Class C2

The tested-on-cooling behaviour was essentially the same as the tested-on-heating behaviour above 2100°F. (1149°C.) but showed less ductility in the range 1800°–2000°F. (982°–1093°C.).

Class C3

For some materials the tested-on-cooling ductility was significantly less than the tested-on-heating ductility for all temperatures.

Table XXI

Behaviour classification of materials hot-ductility tested[140]

Material	*RPI Code*	*On-heating Behaviour*	*On-cooling Behaviour* C (2300)	C (2350)	C (2400)	C (2450)	C (2500)
304	D	H1	—	—	2*a*	—	2*b*
304	M	H1	—	—	2*a*	2*b*	2*b*
304 L	R	H1	—	—	2*b*	—	2*c*
304 L	K	H1	—	—	2*a*	—	2*d*
309	C	H1	—	—	2*a*	2*b*	—
316 Nb	A	H2	—	—	2*b*	—	—
316	E	H1	—	—	3*a*	3*a*	—
316 L	G	H1	—	—	1	2*b*	2*c*
316	L	H1	—	—	2*c*	—	—
347	B	H1	—	—	3*d*	—	—
347	F	H1	—	—	2*b*, 3*a*	2*b*	—
347 Ta	N	H1	—	—	3*d*	—	—
347	O	H1	—	—	3*a*	3*b*	—
347	P	H1	—	—	3*c*	3*d*	—
347	X	H1	—	—	3*c*•	—	—
347	Y	H1	—	—	3*c*†	3*d*	—
347	Z	H1	—	—	3*d*	3*d*	—
347	AA	H1	—	—	3*c*	—	—
347	AL	H1	—	—	3*a*	—	—
347	AS	H1	—	—	1	3*d*	—
N155	AB	H2	—	—	3*d*	—	—
N155	AC	H2	—	—	3*d*	—	—
N155	AD	H2	—	—	3*d*	—	—
N155	AG	H2	1	—	3*d*	—	—
N155	AH	H2	—	—	3*d*	—	—
15–15N	AE	H1	—	—	3*c*	—	—
B2P2	AR	H1	—	—	3*c*	—	—
16–8–2	AT	H1	—	—	—	1	1
16–8–2	BE	H1	—	—	—	1–3*a*	—
Hastelloy B	W	H1	3*a*	3*c*	3*d*	—	—
Hastelloy C	AF	H1	2*a*	2*b*	—	—	—
Inconel X	S	H1	—	1	3*d*	—	—
Inconel	T	H1	—	—	3*b*	3*d*	—
Ni–Zr	U	H2	1	1	1	1	—

• Interpolated data based on behaviour above and below 2300°F.
† No data available below 2300°F. on heating for comparison at lower temperature.
(*a*) Ductility on cooling 60–80% of ductility on heating
(*b*) " " 40–60% " "
(*c*) " " 20–40% " "
(*d*) " " 0–20% " "

It is pointed out by the authors that materials may exhibit class H1 or H2 behaviour on heating and still exhibit C1, C2, or C3 behaviour on cooling. In fact, some materials showed C1 trends on cooling from one temperature and C3 trends when tested on cooling from another temperature.

A summary of the results is given in Table XXI; they lead to a few generalizations about the cracking tendencies of the materials examined:

(i) When a material exhibits H2 behaviour grain-boundary cracking appears likely in joint geometries which supply sufficiently high restraint
(ii) Materials exhibiting H1 behaviour must be examined further
(iii) Materials exhibiting H1 behaviour on heating and C1 or C2 behaviour on cooling should not be susceptible to cracking
(iv) The significance of C2 behaviour is not known at present.

A second means of classification was obtained by measuring the percentage of 'on heating' ductility obtained on cooling. It was found that materials with a known cracking propensity show 'on cooling' less than 40% of their 'on heating' ductility at the same temperature (Table XXI).

The authors claimed that the results of the hot ductility tests appeared to correlate with service behaviour whenever the service was known.

Hot ductility (torsion) tests using a hollow cylindrical specimen with a short gauge length have been recently conducted by Younger and Borland at the BWRA,[161] and some results similar to those of Nippes have been obtained. Some results appear to be significant in relation to susceptibility to base-metal cracking during welding.

Soldan and Mayne,[144] using the same apparatus as Nippes *et al*, made hot ductility measurements on Type 347 material which had been in high-temperature service. These results are recorded in Table XXII and it is evident that there is a good correlation between the temperature ductility pattern and service performance. However, this test is not closely related to service conditions, since the weld thermal cycle follows service conditions instead of preceding them.

Since short-time tests do not, in the main, predict the creep or stress-rupture properties, it would be premature and unwise at this stage to hope that the Nippes hot-tensile test or BWRA torsion test should correlate with service performance. Class C2 behaviour might be significant in relation to service performance and demands much further study.

Parallel work has been published by Puzak and Rischall[141] on the relationship between the temperature of sharp loss of ductility (liquation rupturing temperature) and the solidus temperature. The specimens were raised rapidly to the required temperature, held for two minutes, and then tested in tension. The relationship is given in Table XXIII. The conclusions given are "that these tests demonstrate conclusively that low strength and ductility are developed at temperatures below the bulk solidus temperature for materials determined previously to be of cracking types but not for non-cracking types".

Table XXII

Hot ductility tests correlated with service performance[144]

Material	*Service Life*, hr × 10^3	*Service Performance*	*Red. in Area*, %*
(1) 347 Wrought	22 at 1050°C	Good	72
(2) 347 Wrought	12 at 1050°C	Broke	13
(3) 347 Forged	9 at 1050°C	Broke	19
(4) 347 Wrought	7 at 1100°C	Good	60
(5) 347 Forged	26 at 1100°C	Varied	40
(6) 347 Forged	41 at 1050°C	Cracks	10
(7) 321 Wrought	20 at 1050°C	Cracks	23

* Tested on cooling at 2300°F. (1260°C.)

Table XXIII

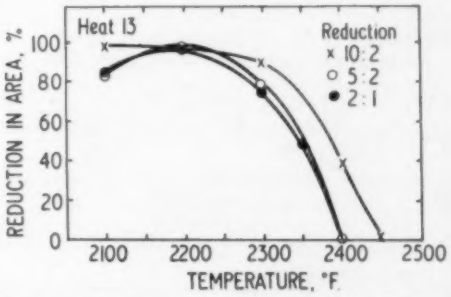
Results of high-temperature tensile tests[141]

Material	*Melt Run Cracking*	*Solidus Temperature*, °C	*Liquation Rupture*, °C
Low C, High Nb 347	Severe	1407	1343
Low C, Low Nb 347	Severe	1407	1371
Standard 347	Moderate	1399	1371
Standard 304	Trace	1393	1400
Low C 304	Trace	1416	1427

Effect of Cold Working and Grain Size

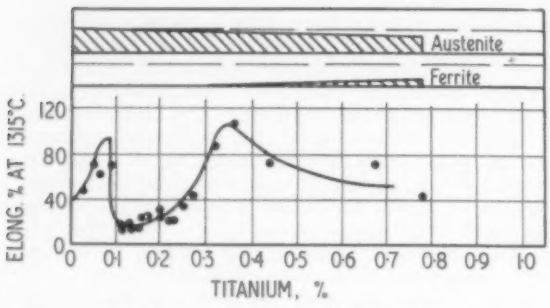
Puzak and Rischall[141] discuss the effect of increased forging reductions (2:1, 2½:1, 5:1) from the ingot on temperature/ductility curves. Results are given (Fig. 25) for low carbon, high niobium 347 material (Nb/C ‖ 33 to 1).

This material was previously[126] stated to be crack sensitive, and it is seen that increased forging reductions have not appreciably altered the liquation rupturing temperatures, thus suggesting little or no redistribution of the fusible segregates. For standard Type 347 (Nb/C=10 to 1) material however, forging reductions to the same extent were adequate to improve the crack resistance. (No temperature/ductility curves are given.)

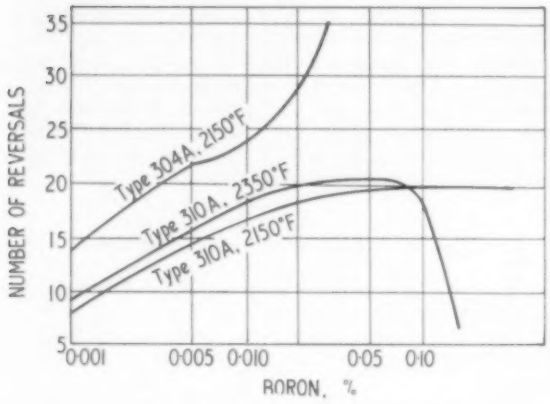


25—*Hot ductility characteristics of low C high Nb Type 347 alloy after various forging reductions from the ingot*[141]

Nippes *et al*,[140] using their hot ductility apparatus, examined the change of behaviour pattern of N–155 alloy (20 Cr–20 Ni–30 Co) after 3:1, 4·5:1, and 9:1 forging reductions. No change in pattern was found from the original (specimens AD, AC, AB, Table XXI).



26—*Effect of titanium on hot ductility*[139]



27—*Effect of boron on hot ductility*[145]

Effect of Alloy Additions

The comments made under this heading are largely concerned with ductility and hot-piercing properties. This subject is important, however, since improvement in the hot ductility of the material is likely to affect its hot cracking and service cracking tendencies.

Perkins and Binder[139] investigated the effect of residual de-oxidation elements. They used short-time tensile tests on specimens of Type 310 material maintained at 1300°C. for 15 min, and recorded the elongation. Their results are summarized as follows:

Si, 0·1–1·5%: A steady increase in ductility from 30 to 100% elongation

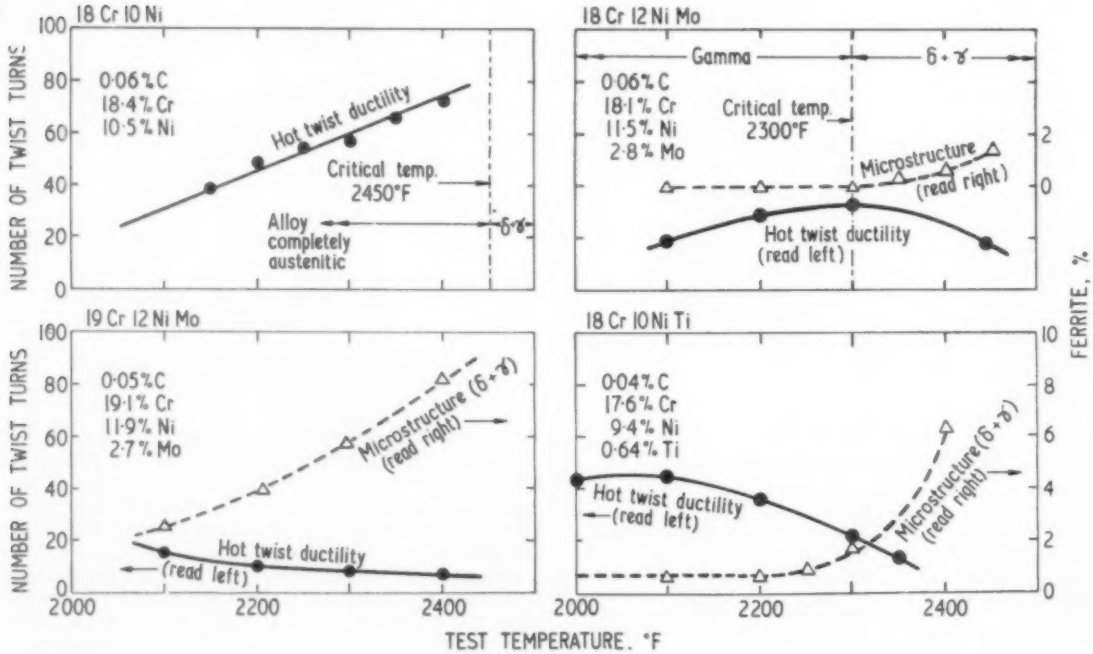
V, 0·2–1·6%: A steady increase in ductility. Ferrite formation beyond 1·6% causes a fall in ductility

Mn, 0·1–5%: A steady increase in ductility. Elongation rising from 30 to 80%

Ti (see Fig. 26): Two ductility maxima are observed. The second coincides with the appearance of ferrite.

Henke and Lula[110] by means of hot (1100°–1320°C.) torsion tests have investigated the effect of rare earths on ductility (number of twist turns to fracture). Although the exact mechanism is not understood, it was found that an increase of 100% in ductility at 1300°C. could be obtained by cerium additions. The optimum amount expressed as % Ce is 0·036–0·040%.

The effect of boron on the number of reversed bends to fracture at temperatures between 1100° and 1200°C. has been studied by Loveless and Bloom.[145] A steady increase of ductility with boron additions was found up to a critical value beyond which the ductility decreased (Fig. 27). This critical amount varies with the Cr, Ni, C, and Mn content of the steel and is



28—*Hot twist ductility*[66]

attributed to the formation of an inter-metallic boride eutectic.

Effect of Delta Ferrite

In investigating the hot-torsion ductility of stainless steels, Bloom, Clarke and Jennings[66] found that austenite and ferrite, when present together, appeared to cause lower ductility than when the predominant phase for a given alloy composition was present by itself. Thus, when the critical temperature ($\alpha \rightarrow \gamma$ or $\gamma \rightarrow \alpha$) of an alloy was passed, the ductility changed sharply, becoming lower if it passed into a two-phase region and superior if it passed into a single-phase region. These authors studied in particular 18 Cr–10 Ni, 18 Cr–12 Ni–3 Mo, 19 Cr–12 Ni–3 Mo and 18 Cr–10 Ni–1 Ti austenitic steels (Fig. 28).

Stress Rupture Measurements

The work of Nippes, Fleischmann *et al*[121,146] is important in this field. In their first paper, the test consisted of a simulated weld cycle on specimens of Type 347 material, followed by stress rupture tests at 40,000 and 45,000 lb/sq.in. at 600°C. In a second series of tests[146] a notched-bar specimen was used with loads from 20,000 to 50,000 lb/sq.in. at 650°C. The results of these tests are shown in Fig. 29. Although a loss of strength was apparent in the 'welded' material the slope of the rupture curve was reduced by the thermal cycle, and the material could not be said to be notch sensitive.

The conclusions of these authors[146] were that high-temperature heat treatment of Type 347 base metal did not impair the mechanical properties of the heat-affected zone, but this does not preclude the possibility of high-temperature liquation causing damage to the structure.

If the materials were heat treated for 2 hr at 1050°C., the original strength of the material was largely restored (Fig. 29).

29—*Stress-rupture strength of high-Ni 347 stainless steel in various heat-treated conditions*[121, 146]

THEORIES OF WELD METAL CRACKING

Since the problem of weld-metal cracking is not restricted to any one type or class of material, it is to be expected that the basic causes of weld-metal cracking are common to all materials. Therefore, a close study of the various reasons put forward for cracking, irrespective of alloy composition, should provide a better understanding of the various factors which are acting to cause, or to prevent, the occurrence of cracks in Cr–Ni austenitic weld metal.

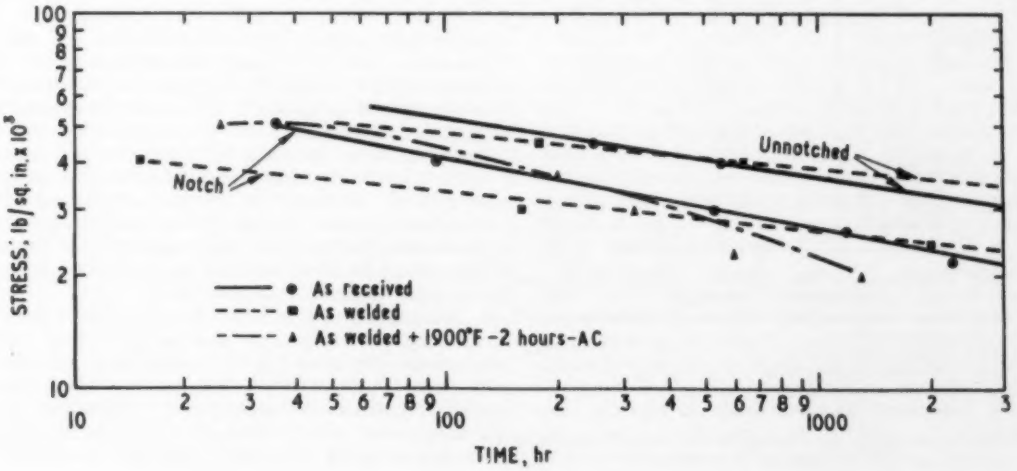
Liquation Theories of Hot Cracking

Many investigators believe that hot cracking occurs at, or slightly above, the solidus temperature. It is argued that the basic cause of cracking is the inability of the semi-rigid network of interlocked dendrites, formed during the liquid–solid stage of solidification, to withstand the contraction strains developed during cooling. Cracks develop where the imposed stresses attain appreciable proportions, and persist if the quantity of liquid is insufficient to heal the fissures thus formed. It is generally agreed that the most dangerous condition exists when there is only a small quantity of liquid remaining at the dendrite boundaries.

Shrinkage : Brittleness Theory

This theory of cracking has been suggested as a result of numerous studies on the 'hot' cracking (tearing) tendencies of aluminium alloys. The origin of the theory probably lies in the work of Vero,[4] but the present-day concept of cracking is due largely to the efforts of Lees,[27,35] Bochvar[7,33,81] and his collaborators, Singer and Cottrell,[24] Singer and Jennings,[32] and Pumphrey and Jennings.[42] The application of the theory to weld-metal cracking is due mainly to Bochvar and his co-workers,[7,33] Pumphrey and Jennings and their collaborators,[24,32,42] Rabkin and Frumen,[54] and more recently to Medovar,[105,118] who has considered the problem of weld metal cracking in stainless steels.

The general mechanism of cracking is thought to be as follows: At some stage in the solidification process



of alloys possessing a freezing range, the primary dendrites, growing at the expense of the decreasing volume of liquid, will come into contact for the first time (coherent temperature) and interlock, thereby forming a coherent network of solid material. During subsequent cooling contraction strains will develop, because the semi-rigid mass is unable to move freely, and when a critical strain (stress) has been exceeded a rift will occur in the solid network. If the rupture occurs when there is a relatively high proportion of liquid present then 'healing' of the cracks can occur. Thus, alloys possessing a high eutectic content are generally free from cracks. Materials of low eutectic content are prone to cracking; if, however, during cooling from the coherent temperature to the solidus, the casting or weld is able to 'accommodate' the contraction strain by means of deforming plastically, cracks will not occur. Cracks are also unable to occur if the dendrites do not come into mutual contact, since no strains are developed, it being assumed that a plentiful supply of liquid is available to fill the spaces between the dendrites.

Strains, in addition to those afforded by the shrinkage of the fused zone, are developed by cooler members of the casting or weld, and thus, in considering the nature of the adverse strains which occur, it is particularly important to pay attention to conditions affording additional restraint.

The brittle range

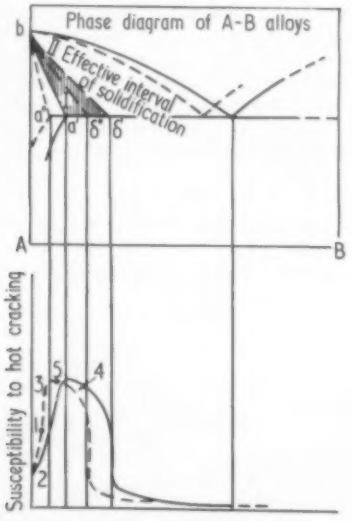
Independent observations by Bochvar and Sviderskaya[33] and by Pumphrey and his co-workers[24, 32, 42] show that the range in which materials are most susceptible to hot cracking is between the temperature at which coherence is gained and the solidus. In Bochvar's terminology[33] this range is the 'effective interval of crystallization': Pumphrey and Jennings[42] use the term 'brittle range'. The latter workers argue that the inherent tendency to cracking above the solidus may be designated the shrinkage-brittleness (σ) and they define this as the contraction per unit volume occurring during cooling through the brittle range. Mathematically this is expressed by the non-dimensional formula:

$$\sigma = 3\,\alpha\theta \qquad \ldots\ldots\ldots\ldots\ldots\ldots \quad (5)$$

where α = average co-efficient of linear contraction over the solidification range
θ = extent of brittle temperature range.

It must be noted that this expression does not take into account the strains imposed by external restraint.

Because of the relationship of the brittle range to the freezing process, the former must be affected by constitutional effects. Thus, in practical conditions of cooling, a depression of the liquidus and solidus temperatures may be brought about by undercooling, while the solidus may be further depressed by lack of diffusion, and in some cases, particularly in steels, by low-melting-point segregates (*e.g.*, iron sulphide in steel). The crystal size and habit largely influences the degree of liquid entrapment, a high degree of entrapment being favoured by long and slender dendritic formations. The dendritic formation is in turn promoted by rapid rates of linear crystal growth, and only small numbers of crystal nuclei. Entrapment is offset by flat temperature gradients, and excess nuclei which promote equiaxial grain growth.



b–δ′ Initial temp. of formation of continuous crystal network under equilibrium conditions
b–δ″ Same under actual conditions at crystallization of molten pool
b–α′ Equilibrium solidus
b–α″ Dis-equilibrium solidus
I Region of solid/liquid state
II Region of liquid/solid state

30—*Correlation among hot cracking (hot shortness), quantity of eutectic and 'effective interval' of solidification*[105]

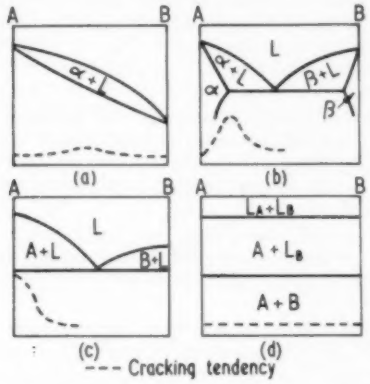
Relationship of Constitution Diagrams to Cracking

The first realisation that constitutional features were significant in relation to hot-cracking tendencies appears to occur in the work of Vero.[4] Present-day theories are due to Bochvar and his co-workers,[33, 81] Pumphrey and his school,[24, 32, 42] and Medovar.[118]

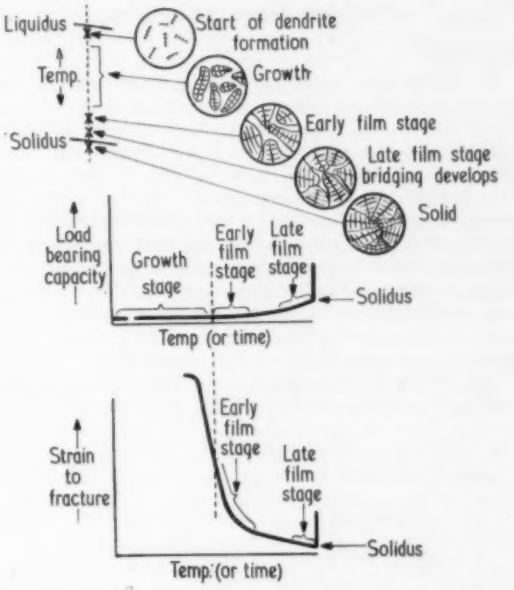
First consider the eutectiferous system (Fig. 30). From what has already been stated, it is apparent that the alloys most susceptible to cracking are those which have compositions with the broadest brittle range. Referring to Fig. 30 these alloys have compositions corresponding to the limit of solid solubility. Compositions with lower or higher alloy contents are less prone to cracking. In the former case the reduced cracking is due to a smaller quantity of liquid combined with the reduced brittle range. For compositions to the right of the point of solid solubility the cracking tendency decreases with increasing alloy content, owing to large quantities of liquid (eutectic healing) and grain structure refinement. The critical proportion of eutectic necessary to prevent cracks has been estimated for a few aluminium alloy systems, but not for ferrous alloys. It has been suggested, for instance, that an Al–Si alloy containing more than 20% by volume of liquid does not crack.[32]

The explanation of the effect of constitution on cracking tendency for the eutectiferous system described (see also Fig. 31*b*) has been extended by Medovar[118] to include other simple binary systems (Fig. 31). Thus, for example, the variation of hot cracking with alloy additions, for elements forming continuous solid solutions, would be determined by

the length of the freezing range (Fig. 31*a*). For systems that do not form solid solutions at all it is to be expected that even small additions of the second element would cause intense cracking owing to the rapid increase in the extent of the brittle range. Further additions should tend to reduce cracking (Fig. 31*c*).



31—*Basic types of binary diagrams and cracking tendencies*:[118] (a) *Continuous solid solutions*, (b) *restricted solid solution*, (c) *no solid solution*, (d) *elements mutually insoluble in liquid and solid state*



32—*Mechanism of hot cracking as deduced from casting studies*[103]

If the elements are not mutually soluble in either the liquid or solid state then hot cracking should not occur (Fig. 31*d*). It is argued that the liquid would be squeezed out by the growth of the crystals, and rise to the surface of the weld.

In general, if the freezing range is very short, then hot cracking should not occur.

Strain Theory of Hot Tearing

The strain theory of hot tearing advanced by Pellini and his co-workers[75,89,90,129] suggests that the prime cause of hot cracking in welds and castings is the localized strains set up by thermal gradients tending to pull apart solid masses of material separated only by continuous thin films of liquid (Fig. 32). This concept is in fact (but not in detail) similar to the shrinkage-brittleness theory. The strain theory, however, only considers the case of thin films existing at near-solidus temperatures, and does not consider the more general case of the brittle range concept. In fact, it is suggested that hot tearing cannot occur during the mushy stage (unless the strain rates are extraordinarily high) since the shrinkage strain of the hot zone is distributed relatively uniformly. When the film stage is reached, the total strain is localized to a few film zones and fracture is able to occur. This concept of cracking is, however, too specific. It may be argued in many circumstances that the severity of cracking depends on the temperature of crack initiation, and that cracks initiated at temperatures close to the coherent temperature would be more severe than cracks started very close to the solidus, where only very thin liquid films exist. In the latter case the fissures would be very small and numerous. Since both large cracks and fissure-type defects are found in practice, it seems unlikely that the strain theory (as proposed in its present form) is valid for both types of hot crack. In agreement with the shrinkage-brittleness theory, the strain theory concept indicates that the cracking susceptibility is increased by an increase in the solidification time.

Other Theories

Whether or not micro-fissures observed during high-temperature service are present immediately after welding, or whether they are initiated during the service period (stress-rupture failures) is still a very debatable point.

Some authors believe that cracks occur in the subsolidus region, it being understood that the existence of liquid films is not a necessary feature for crack initiation.[107] Carpenter and Jessen[40] believe that intergranular silicate films are the prime cause of cracking. Heuschkel[125] believes that, in some cases, fissures occur during the weld thermal cycle, but also that many are formed by the stresses imposed upon the weld joint when the material is subjected to high-temperature service conditions. The reason for this view is that many Cr–Ni austenitic steels have a pronounced dip in the temperature/ductility curve at about 1800°F. (982°C.).

Heuschkel[125] shows that the high-temperature brittleness is related to the strength of the matrix; the higher strength matrix imposes high stresses on the grain boundaries, which, if contaminated, will separate under low orders of strain. Grain-boundary weakness is related to the presence of oxides, nitrides, carbides, and silicates at the grain, sub-grain, and particle boundaries, even when the sulphur and phosphorus contents are kept as low as 0·015%.

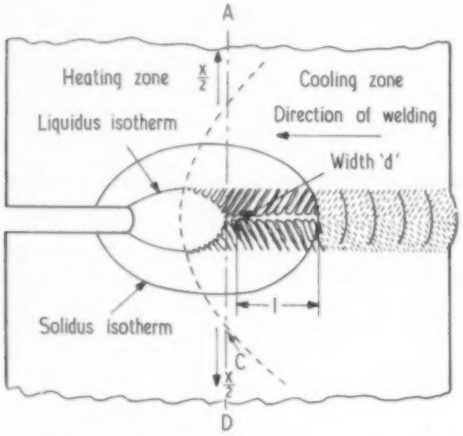
Particular Forms of Cracking

Longitudinal

In considering hot cracking in a weld between two parallel sheets of aluminium alloy, Pumphrey and Jennings,[42] using the shrinkage-brittleness concept, approached the problem from a fundamental viewpoint. Using the model shown in Fig. 33 they argued that longitudinal cracking arises from the inability of liquid metal to flow freely and compensate for the contraction of both liquid and solid parts (including the parent plate) of the weld joint. Cracking is thus more likely to occur by an increase in x, l or d. An increase in x is effected by an increased external restraint; l is increased by an increase in the thermal expansion, the speed of welding, and the brittle temperature range.

The effect of strain conditions imposed on the weld during the fusion process has been considered by Terry and Tyler.[154] Thus in Fig. 33 the total contraction which causes longitudinal cracking comprises contraction in zone BC (cooling region) minus the expansion in zones AB and CD (heating region). No cracking occurs if the total contraction is zero. Increasing the welding speed, thus making the contraction zone smaller, should reduce cracking.

The hypothesis that cracking should be reduced by increasing the welding speed is not in accord with Pumphrey and Jennings' suggestion.



33—*Conditions obtaining in a butt weld between two parallel and restrained metal sheets*[42, 154]

Transverse

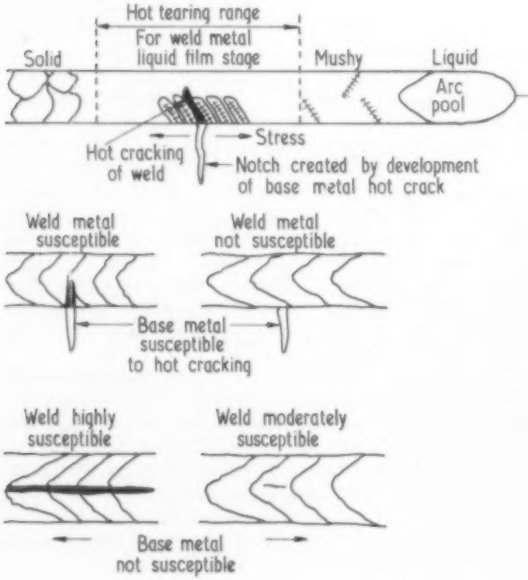
Hot cracking in stainless steel weldments has been discussed by Pellini[90] and his associates[103,126,141] by considering the strain theory of hot tearing. It is considered that base metal cracks could 'trigger off' cracks in the weld metal.

The necessary conditions for weld-metal cracking are the presence of liquid films (low-melting-point segregates) (Fig. 32) which persist during the solidification period.

The mechanism of cracking is depicted schematically in Fig. 34. Cracking is developed in weld metal only when a thin film condition exists in the solidified weld bead. Cracks do not develop during the mushy or pasty stage of solidification because of the high proportion of liquid present.

THEORIES OF BASE METAL CRACKING

Cracks visible immediately after welding are thought to be due either to liquation or high-temperature brittleness. When cracking has occurred during stress relief or high-temperature service then stress-rupture theories of failure are proposed. These latter theories sometimes recognize the fact that micro fissures were present in the metal before post-weld treatment was begun.



34—*Schematic representation of weld metal hot cracking*[126]

Liquation Theories of Cracking

Origin and nature of grain boundary liquid films

Pure Alloy

Values for the grain boundary melting point for a pure metal are controversial, but estimates of the difference from the bulk solidus are confined to from 0·1° to 2°C. In an alloy, a false solidus may exist in the casting solidification and this coring effect may be retained if subsequent forging is insufficient. In a heavily cored structure (Pb–Sb system) the solidus value for the last liquid to freeze may be up to 50°C. below the true solidus value.[90]

Impure Alloy

The presence of low-melting-point segregates in the grain boundary is a further source of grain-boundary

liquid films. The origin of this segregate could be in casting, where the subsequent rolling or forging had failed to break up the films, or could be due to solid state diffusion.

When materials are hot worked above the solidus, they become hot short only if the liquid phase has a small dihedral angle and so replaces a large part of the grain boundary area. In welding, this phenomenon could be important to base-metal cracking tendencies.

The distribution and continuity of these liquid films is dependent on the dihedral angle of contact[43] (Fig. 35) between the liquid and the solid grains. It may also be shown that the dihedral angle is a function of the interfacial energies of the grain boundary and of the solid/liquid interfaces. Thus:

$$\gamma LS \cdot \cos \theta/2 = \tfrac{1}{2}\gamma SS$$

where θ = dihedral angle.
γLS = interfacial energy, film/solid, ergs/sq.cm
γSS = interfacial energy solid/solid (grain-boundary), ergs/sq.cm

Variables which alter the $\gamma SS/\gamma LS$ ratio will alter θ. Among these variables are:

(*a*) The chemical composition of the film
(*b*) Temperature. This may alter the chemical composition of the film, and may also affect the percentage of delta ferrite and decrease the number of austenite/austenite grain boundaries. These have a significantly higher energy than γ/α or α/α grain-boundary interfaces.

If $\frac{\gamma LS}{\gamma SS} = \infty (\theta = 180)$ the segregate film is confined to the triple grain points, and parting of the grains by stress is not generally facilitated.

If $\gamma LS = \frac{1}{\sqrt{3}}\gamma SS$ $(\theta = 60^\circ)$ the segregate is continuous round the grain edges, and between $\theta = 60^\circ$ and $\theta = 0^\circ$ notch effects and stress raising may facilitate parting.

If $\gamma LS \leq \frac{1}{2}\gamma SS$ $(\theta = 0)$ the segregate envelopes the grain entirely and any imposition of stress may result in grain separation. Liquid is able to penetrate the grain boundaries only when $\gamma LS < \frac{1}{2}\gamma SS$.

The application of stress can, according to McLean[127] displace the condition for grain-boundary penetration by the liquid. To make an estimate of the magnitude of this effect, suppose that the atomic cohesive forces extend over three atomic diameters. Then the work done per sq.cm of boundary by the stress during intergranular fracture is $3Fb$, where F is the applied stress and b the atomic diameter. The modified condition for penetration is then

$$2\gamma LS \leq \gamma SS + 3Fb$$

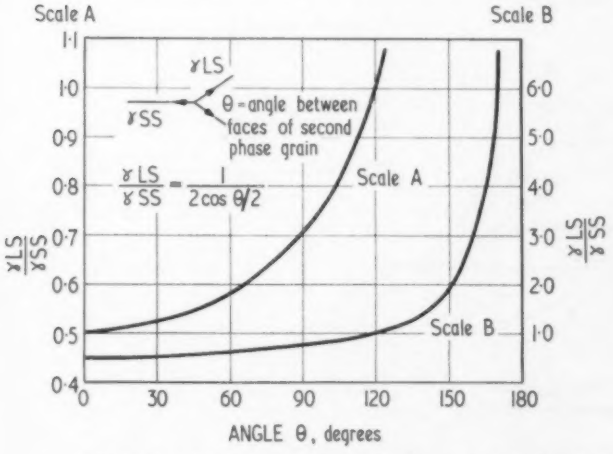
Liquation due to high-temperature ferrite

The role played by high-temperature ferrite in segregation and liquation has been discussed by Parks[124] in application to weld metal but the theory could equally be applied to base metal. Thus in a typical thermal cycle (shown diagrammatically in Fig. 36) delta ferrite is stable for a time interval t_1, when temperature is high and diffusion rates are rapid. According to Parks, this should lead to diffusion into the ferrite of such alpha stabilizers as Si, P, Cr, Mo, W, V, Nb, Ta, Ti, and Zr, whose free energies of solution are lower in ferrite than in austenite. During the time interval t_2 the temperature is lower and diffusion is restricted, so that as the ferrite transforms there is a gradual concentration of these stabilizers in the ferrite lattice. This may lead to liquation of the ferrite/austenite interface during cooling or during reheating owing to the concentration of certain elements above solidus values (*e.g.*, Si and P).

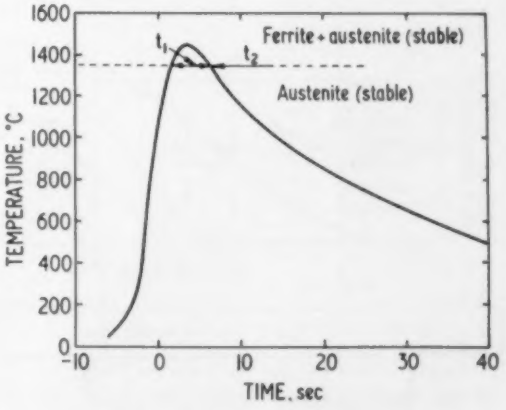
Cracking due to pre-existent liquid films

Base Metal

This mechanism has been discussed by Puzak, Apblett, and Pellini[126] as an application of the latter's 'Strain Theory of Hot Tearing'[90] and is illustrated schematically in Fig. 37. On welding, the base metal adjoining the weld is stressed initially in compression owing to the thermal expansion of the hot metal near the fusion line. Upsetting is developed at this stage. As the welding arc passes on, tensile stresses are set up and these result in the separation of the liquid film region, thus producing a base-metal crack. The position and relative extent of cracking is dependent on the amount of liquid film present, the unit strain rate



35—*Ratio of interphase boundary tension and grain boundary tension as a function of dihedral angle of second phase*[43]



36—*Typical thermal cycle in heat-affected zone of 1½ in. stainless steel plate*

developed, and the film life. These concepts are illustrated in Fig. 32, which shows that contraction strains of the whole work piece are concentrated on the liquid films, and if these have a low melting point (long film life) the contraction strain is too high for cracking to be prevented.

The eutectic healing mechanism is inapplicable to base-metal cracking, since there is no liquid source present which is not under tension stress. Another factor which is important is the difference in temperature at which the base metal and weld metal solidify respectively. For stainless steel the solidus temperature of the weld metal is usually similar to that of the base metal, but this is not always so. Until the weld metal attains coherency, it is not capable of transmitting stress to the base metal and, since by this time the base metal may be below its solidus temperature, it will be able to withstand this stress. If, however, the solidus of the weld metal is above that of the base metal, the weld may attain coherency and transmit additional stress to the base metal while liquid films still exist. This may result in aggravated base-metal cracking.

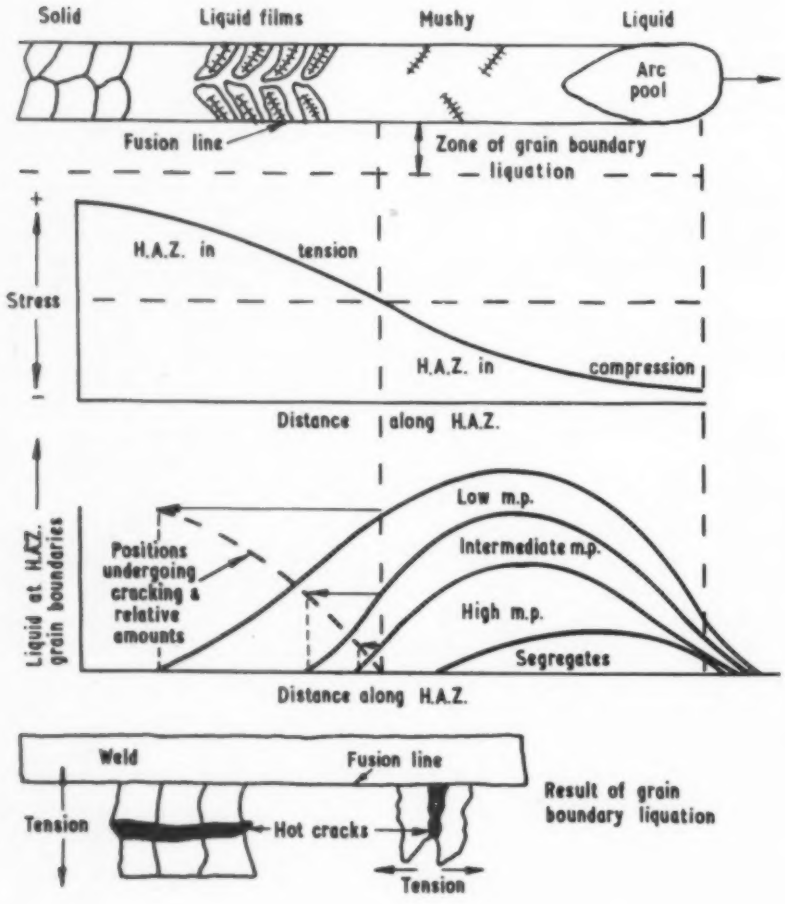
Fusion Line

This mechanism is again discussed by Puzak *et al*[126] but no differentiation from base-metal cracking is made (Fig. 37). The important difference, however, lies in the existence of a source of liquid metal in the fusion zone which can cause healing of the base metal. If this base metal is in a state of tension parallel to the weld and liquid films exist, then hot tearing will occur, but if also the weld metal is still above a critical temperature, liquid healing will take place. This mechanism relies on a favourable interval between weld coherency and base-metal solidus.

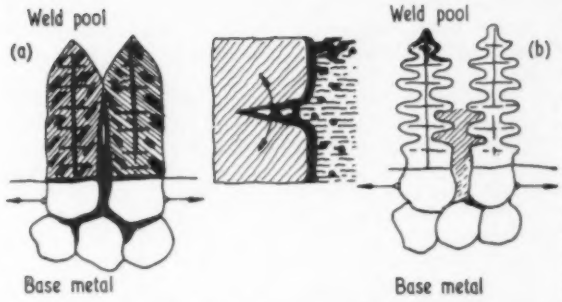
The second important difference between base-metal and fusion-line cracking lies in the theory (Apblett and Pellini[103]) that the latter is capable of initiating weld-metal cracking by a notch extension mechanism. This has already been discussed.

Cracking tendency of fusion line due to Rehbinder-Medovar effect

A different mechanism to explain the existence of continuous cracks from weld metal to base metal is provided by Medovar.[147] The theory is based on the earlier work of Rehbinder[36,41] who suggests that adsorbtive migration of certain liquid substances into metals may occur. By the Rehbinder effect certain



37—*Schematic representation of base metal hot cracking*[126]



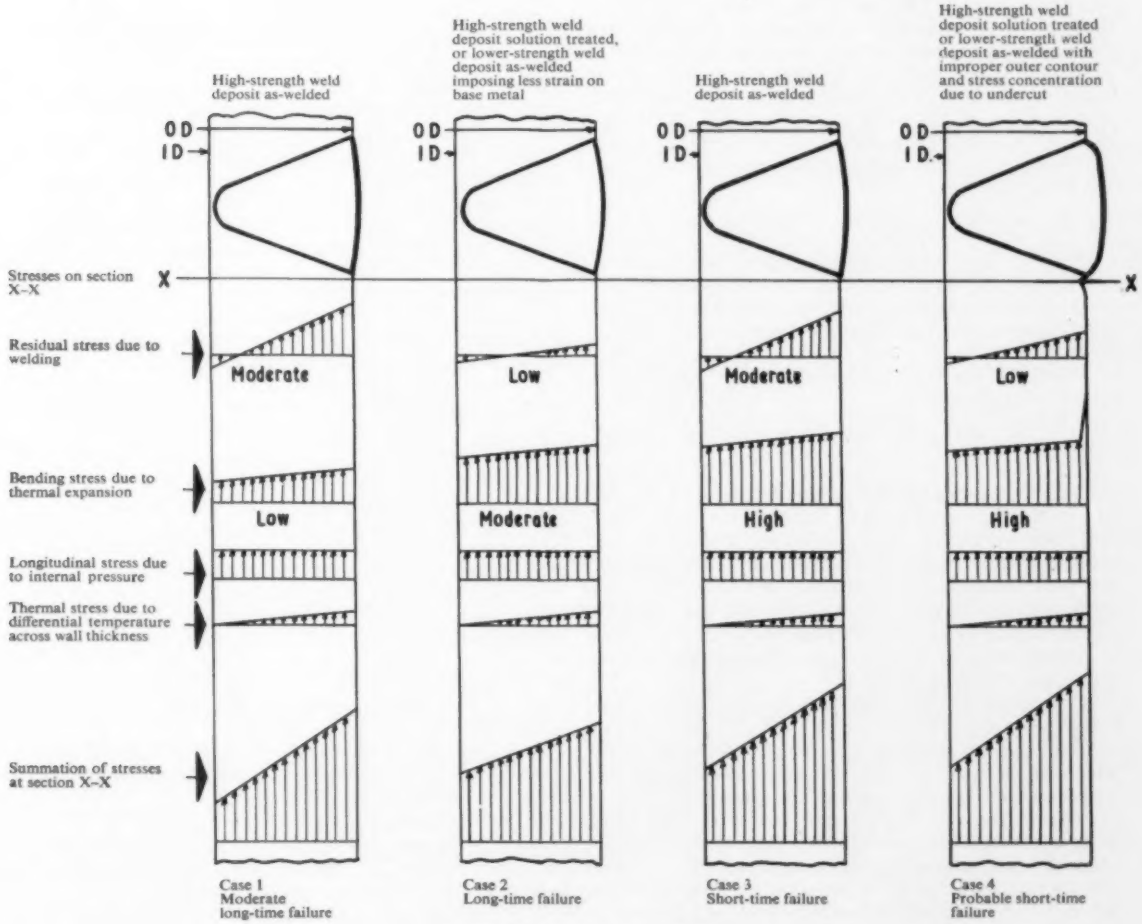
38—*Schematic formation of intercrystalline heat-affected-zone cracks as a result of the Rehbinder absorption effect*:[147] (a) *Crack not refilled with liquid*, (b) *crevice between parent metal grains refilled (healed) with low metal parent liquid*

substances which are capable of lowering the surface energy of a solvent ('surface active' elements) are capable of dynamic penetration of a metal by adsorptive migration. Such elements in liquid steel are given[25] as B, Cu, Mo, W, Ti, Al, Si, Va, Cr. These elements are preferentially adsorbed at the weld-metal/base-metal interface, especially where the weld metal film stage has further concentrated them. These elements may then penetrate the grain boundaries, lowering the solid/liquid interfacial energy and forming a continuous base-metal liquid film. Under the influence of tensile forces, strain will now develop, and (assuming that healing is prevented) cracking will occur. The mechanism is presented schematically in Fig. 38. The importance of this mechanism is that a base metal completely free from fusible intergranular segregates is still capable of cracking if the weld metal composition is unsuitable.

The McLean[127] condition for grain-boundary penetration is obviously a logical extension of Rehbinder's ideas.

Stress Rupture Theories of Cracking

As has been stated previously, stress-rupture theories are only applicable to cracking in stress relief



39—*Weld stresses in service*[143]

or in service. It is possible that cracks initiated during welding are propagated during stress relieving or in service.

Emerson and Jackson[142] state that the predominant factors responsible for these failures are believed to be high levels of combined stress and the inability of the material to absorb the imposed stress by plastic deformation at the operating temperature. In heavy-wall pipe joints values of residual stress of 25,000 to 35,000 lb/sq.in. are quite common. These authors state that stress values are additive and include the following (Fig. 39):

(i) Residual stress due to welding
(ii) Bending stress due to thermal expansion
(iii) Longitudinal stress due to internal pressure
(iv) Thermal stress due to differential temperature across the weld thickness.

High localized stresses are possible from notch effects resulting from arc strikes, undercut, etc., and it is postulated that in this way the rupture stress is exceeded.

Poole[98] states that cracking during stress relief might be a stress rupture type of failure. To avoid excessive stresses from thermal gradients in heavy sections, slow heating rates are usually specified, and consequently the time during which a highly stressed weldment is at a given temperature may exceed the stress rupture life of the metal. One of the strongest points in favour of a stress-rupture mechanism is that, with one exception, all cracking reported during stress relief occurred when heating and/or cooling rates were less than 110°C./hr.

CONCLUSIONS

Cracking in Cr–Ni austenitic weldments can occur during any one of the following stages:

(*a*) Weld thermal cycle (visible immediately after welding)
(*b*) Stress relieving
(*c*) Service.

Cracks formed during Welding

Weld metal

Weld metal cracks which occur during welding of the fully austenitic materials are probably, in most instances, due to the presence of small quantities of liquid residing at the grain boundaries while the weld is solidifying and, at the same time, being subjected to moderate or excessive strain.

Elements which are extremely harmful, even when present in very small amounts (less than 0·1%), are sulphur and phosphorus. Less harmful elements, but which still demand careful control are carbon, silicon, niobium, zirconium(?), copper and vanadium(?).

If these harmful elements are present in relatively large quantities then the probability of cracking is reduced because of the reduced strains developed, the beneficial effect of grain refinement, and the fact that the increased quantity of liquid (eutectic) acts as a filler. Thus, referring to Fig. 30 (eutectiferous system), cracking reaches a maximum at the maximum extent of the brittle range and, thereafter, decreases. Weld metal of near-eutectic composition is ideal from the weldability viewpoint.

The presence of 5–10% ferrite in the weld structure (austenitic/ferritic alloy) prevents cracking; smaller quantities will frequently prove effective. Its beneficial action is probably due to its capacity for dissolving larger quantities of harmful elements like phosphorus, silicon, and niobium, than can austenite, and to its action in creating an increased grain-boundary area, which thins out the liquid films and also lessens the tendency to segregation. The Parks' hypothesis that ferrite acts to concentrate the ferrite formers and cause liquation demands further study before it can be accepted.

High-temperature brittleness is another factor which has been proposed as a cause of cracking, but it seems likely that this low-ductility region (about 1000°C.) would not be troublesome, because of the limited temperature range in which it is found. This factor is probably important in post-weld heat treatment or in service, or both.

Base metal

Base-metal cracks not connected to the fusion line are probably due to the presence of liquid films of lower melting point than the bulk material. Liquid healing is unable to take place. Although the liquid films cannot be modified in character by additions to the weld pool (*e.g.*, ferritizers) cracking may be reduced by altering the composition of the weld metal to decrease the strain. It should be possible to deposit weld metal which is coherent only at a temperature lower than the solidus of the base metal, and thereby reduce the strains imposed upon the base metal during its liquid film stage. Equally well, a weld metal with lower plastic resistance than that of the base metal (*e.g.*, 16Cr – 8Ni – 2Mo on 18Cr – 12Ni – 1Nb steel) should be beneficial, since the strains on the base metal should be lessened.

Fusion-line cracks can be started by low-melting-point segregates in the base metal and once formed can, under unfavourable circumstances, propagate into the weld metal.

Fusion-line cracks may also be caused by the Rehbinder-Medovar effect (grain-boundary penetration). In this case cracking may be reduced by modifying the composition of the weld pool. Copper appears to be the element which is most harmful in this respect.

High-temperature brittleness has also been proposed as a cause of cracking, but supporting evidence is lacking.

Cracks formed after Welding

It is not clear from the literature whether the initiation of service failures takes place during welding, during stress relief, or during service.

It has been suggested that the properties of the heat-affected zone are unimpaired by the welding cycle, and that cracking is due to the high stress levels existing in the weldment.

This suggestion, however, is not universally accepted and requires further study.

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BOOK REVIEWS

"*Welding Handbook*", Sect. 2: Welding Processes—Gas, Arc, and Resistance. 4th ed.: American Welding Soc., 1958, Cleaver–Hume Press, 80 pp. (72*s*.)

The Handbook of the American Welding Society has long been accepted as the most complete compendium of welding information and its successive editions have grown to a considerable volume with the growth of welding practice and the development of new processes.

The fourth edition now appearing is in five separate annual volumes; Section 2, just published in Great Britain, covers the welding processes—gas, arc and resistance. Much of its value stems from the fact that the information is compiled by teams of experts from industry; it is thus complete in the fields covered and does not represent the rather narrower views that might come from an individual editor.

A good deal of the content of each Section is virtually the same as in the previous edition, and this is natural in a handbook of this type which covers welding from fundamentals, but up-to-date information has been added on the technological advances made since the last edition was published in 1950, particularly on the more recently developed automatic processes.

More information is now included on inert-gas metal-arc welding and on submerged-arc welding, including methods of welding with more than one wire. The Section on arc welding plant is expanded to cover rectifiers, constant voltage and rising voltage machines, and the Section on resistance welding now includes details of the newer control systems and adaptations of the processes.

This is an American publication and does not include some welding processes used in Europe, and the terminology, tables of welding symbols, etc., are different from European practice, but these are small criticisms of a well-established and most useful encylopaedia of welding information.

The format and layout are greatly improved and the volume is well indexed with bibliographies for further sources of information at the end of each Section.

R. BUTLER

D. SÉFÉRIAN, "*Metallurgie de la Soudure*", Paris, 1959, Dunod, 393 pp.

The first part of this book, dealing with the metallurgical phenomena which occur in the welding of steels, describes the various welding processes, expounds the iron-carbon equilibrium and outlines the psuedo-equilibriums which can result from the interaction of the metal deposited by the arc and the electrode coverings.

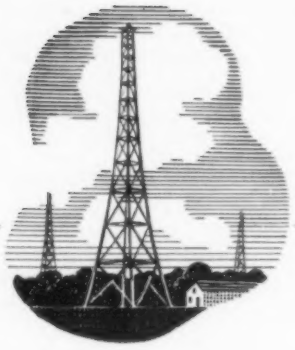
One chapter deals with the fundamental problems of cracking. The preheating temperatures are set out and compared with the results of experiments in Britain.

In the second part of the book there is a study of the weldability of many alloy steels, such as the nickel, chromium and molybdenum steels, as well as of the newer steels with boron, and the so-called semi-refractory steels with chromium and molybdenum, used in the power industry and the oil industry.

The book treats welding problems at a fully scientific level, and is intended not noly for design engineers, fabricating engineers, and welding technicians, but also for research workers and engineers in the steel industry, who must take account of the weldability properties of different steels.

P. D. BOYD

NEWS
and Announcements



News of the Institute and Branches
B.W.R.A. · Other Societies ·
and Industry

INSTITUTE ACTIVITIES

Personnel of Council

The following, on election as Chairmen of Standing Committees, have become *ex officio* members of Council for 1959/60: Professor J. G. Ball, Mr. R. G. Braithwaite, Mr. J. F. Lancaster, Mr. E. Fuchs, Mr. J. A. Dorrat.

At its meeting of 19th November, the Council accepted the resignation of Mr. E. Fuchs as one of the representatives of Fellows, Members and Associate Members, and resolved to co-opt Mr. E. Flintham for the remainder of the year.

The appointment of two Branch representatives was reported at the same meeting, namely, Mr. J. E. Roberts (West of Scotland Branch) in succession to Mr. D. M. Kerr, resigned, and Mr. I. B. Ford (South Wales Branch).

Tibbenham Prize

The Tibbenham Prize, endowed by Mr. F. J. Tibbenham, which is awarded by the Council of the Institute on the results of the Welding Certificate Examination of the City and Guilds of London Institute, has been awarded to Mr. **B. E. J. Hanling** of the Lancaster and Morecambe College of Further Education.

"Health and Safety in Welding"

At its November meeting, the Council approved in principle proposals for the publication of a revised edition of the Institute's booklet on Health and Safety in Welding, to incorporate a new chapter on safety requirements for certain allied processes. Over 20,000 copies of the first edition of this booklet were sold.

Projectionist

The Institute has need of the services of a projectionist to operate the epidiascope and the film projector during courses in the School of Welding Technology and at evening meetings.

The School courses are held in alternate weeks, and normally occupy the mornings and afternoons of five days. The evening meetings are two or three a month during the winter, and last from 6.0 to 9.0 p.m. It is thought that some members of the Institute living in London might be interested in giving part-time service in return for an honorarium. Anyone interested should write to the Secretary for further particulars, explaining how much time he would be prepared to give.

MEETINGS

Lecture to Young People

Professor Hugh O'Neill is giving the Institute's first Christmas Lecture to young people at the Institute on Tuesday **5th January** at 11.00 a.m. His title is "*The Magic Arc*", and the subject is, of course, the use of electric arc welding. Demonstrations will be given. Admission will be by ticket only, and members are invited to apply for their own children, or for friends', between the ages of 15 and 18.

Conversazione: IIW work in hand

On **4th February**, from 4.0 to 8.0 p.m., there will be a conversazione at the Institute, with a number of exhibits illustrating the technical work of the International Institute of Welding and the part which the United Kingdom Delegation is playing in it. Admission will be by invitation, and members who are interested and would like to have a ticket are invited to apply to the Secretary.

Third Annual Lecture

The Institute's Annual Lecture will be given by Mr. W. G. John, Deputy Director of Naval Construction, Admiralty, at the Institute on the evening of **Thursday, 3rd March.** The subject will be the influence of welding on naval architecture.

Spring Meeting, Droitwich, May 1960

The Spring Meeting of the Institute will be held at Droitwich from **9th to 11th May 1960,** and will be devoted to the novel welding processes and metal spraying. The provisional programme includes an opening reception at the Worcestershire Brine Baths Hotel in the early evening of Monday, 9th May, two morning sessions, on Tuesday and Wednesday, for the discussion of technical papers, demonstrations on the Tuesday afternoon and early evening, a choice of works visits on the Wednesday afternoon and a concluding dinner that evening.

The conference centre will be the Winter Gardens, Droitwich, and most of the supply manufacturers will be taking part in a small exhibition of photographs, literature, etc.

Five papers on different aspects of metal spraying have already been promised, and it is intended to devote the whole of one technical session to these.

AUTUMN MEETING, 1959

The Autumn Meeting of 1959 was a joint one with the Société des Ingénieurs Soudeurs of France, nearly fifty members of which, headed by their President, Monsieur L. de Jessey, their Secretary, Mademoiselle L. Blosset, and Monsieur A. Leroy, Director of the Institut de Soudure, attended some part of the proceedings.

The French visitors were greatly impressed by the meeting, and the Institute has received from them many congratulatory letters and the warmest expressions of appreciation for the welcome accorded to them.

A party of members of the French Society of Welding Engineers at the Crawley factory of the A.P.V. Company. They are seen here with Dr. Richard Seligman, A.P.V.'s Founder and President, who welcomed them before they started their tour of the Company's extensive engineering works, fabrication shops, foundries, and laboratories

The bulk of the French party arrived on the night of Monday, 2nd November, and joined one or other of the two works visits organized for them on the following day. One party went to the British Welding Research Association at Abington, where they were welcomed by the Director of Research, Dr. R. Weck, and his staff, and were conducted through the laboratories after luncheon. The other party went to lunch at the George Hotel, Crawley, where they were met by Mr. M. J. P. Martin of the A.P.V. Company Limited, and were then taken on a tour of the works, ending with tea before they returned to London.

Meantime, at Princes Gate, the President, Mr. E. Seymour-Semper, delivered his Presidential Address to members, after a Press Conference on Welding Education. The President then held a reception which the French visitors joined on their return from Abington and Crawley. After the Reception, Mr. and Mrs. Seymour-Semper entertained the President and other officers of the Société des Ingénieurs Soudeurs with dinner at the Goring Hotel.

On the next day the French visitors again broke into two parties, one visiting the works of G. A. Harvey & Co. Ltd., at Greenwich, and the other the research laboratories of Murex Welding Processes Ltd. Mr. E. F. Burford at Harvey's and Mr. J. M. Willey at Murex, with members of their staffs, welcomed the visitors, for whom most interesting tours had been arranged. After lunch as the

At the Annual Dinner

Mr. John Strong, Past-President (left), talking with Mr. A. C. Jennings, Master of the Worshipful Company of Founders, behind (centre) Mr. Edgar Fuchs, Vice-President

Professor O. A. Saunders (left) with Sir Andrew McCance

Monsieur L. de Jessey, President of the French Society of Welding Engineers (centre) in conversation with Lord Harvey of Tasburgh (left) and Professor Cyrille Arnavon

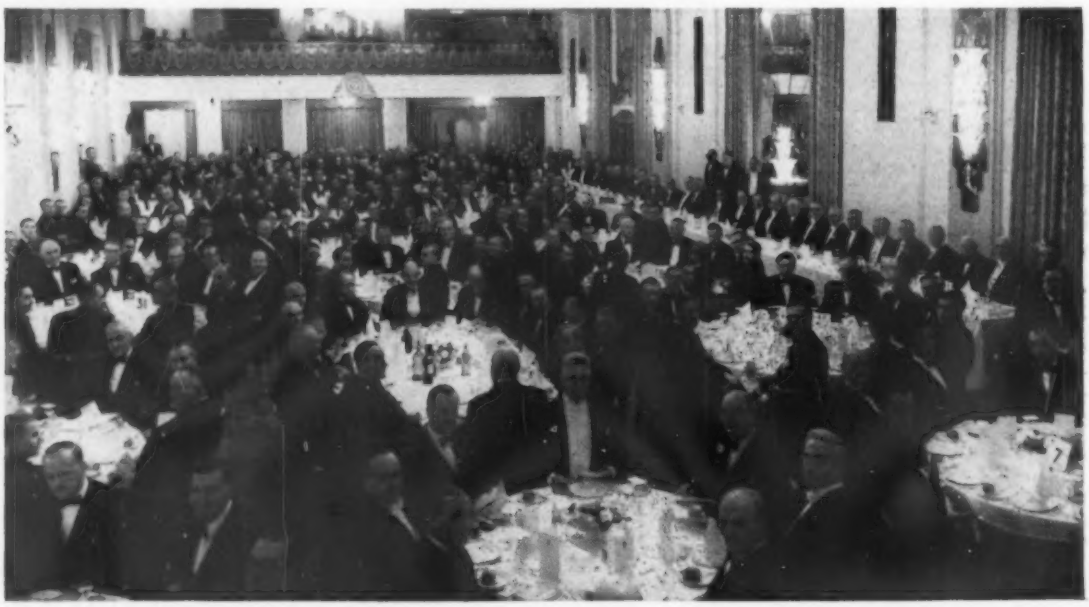
At the Annual Dinner

Monsieur A. Leroy (left), Director of the Institut de Soudure, talks with Mr. P. L. Taylor, Secretary of the British Acetylene Association

guests of the two companies, the visitors returned to their hotels in the early afternoon.

That evening the Annual Dinner of the Institute was held at the Park Lane Hotel with the President in the Chair, and a large number of distinguished guests, headed by Sir Edward Boyle, Bt., M.P. (Financial Secretary to the Treasury), and the Rt. Hon. Lord Harvey of Tasburgh (Chairman of the Franco-British Society). The order of the Toasts had to be reversed, because Sir Edward Boyle was detained at the House of Commons by parliamentary business, and the first speaker was, therefore, Mr. J. Strong, immediate Past-President of the Institute, who proposed the Toast of "The Guests" in a speech which charmed and delighted all his hearers. Lord Harvey of Tasburgh and Professor O. A. Saunders, President-Elect at the Institution of Mechanical Engineers, responded. Sir Edward Boyle then proposed the Toast of the Institute and the Société, paying particular tribute to the pioneering work of the French Institute in the education of welding engineers, and stressing the importance of the place which welding must occupy in the Government plans for the expansion of technological education in Great Britain. The Presidents of the two bodies then responded, Monsieur de Jessey speaking in French and in English.

The Institute's Lecture Hall was crowded for the opening of the Technical Session on Thursday morning, when four papers were presented and



Members and guests at the Annual Dinner at the Park Lane Hotel

discussed, under the chairmanship of Mr. E. Fuchs, Vice-President. The papers, three by French authors and one by the Chairman and Mr. W. Ashworth, were as follows:

"Application of Standard Weld Seams to Boilermaking" by *P. Cahany and H. Gerbeaux*
"Examples of the Organization of an Inspection Department covering Ships, Boilers and Welded Frames" by *P. Pouchet, J. Hennion, J. Delair and J. Le Burn*
"Quality Control in the Construction of Pipelines for the Transport and Distribution of Petroleum Products and Combustible Gases" by *H. Gerbeaux*
"Quality Control Methods adopted by a Large User when Purchasing Welded Equipment" by *E. Fuchs and W. Ashworth*

Eleven separate speakers had given notice of their wish to take part in the discussion, but, in spite of the excellent interpretation by Mrs. Saxe, it was impossible in the time available for them all to speak or for all the authors to reply, so that the Chairman had to invite written contributions.

The afternoon meeting was even more overburdened with five papers and seventeen speakers in the discussion, and again written contributions had to be invited, despite the prolonging of the meeting until nearly 7 o'clock. The Chairman was the President of the Société des Ingénieurs Soudeurs, Monsieur L. de Jessey. The papers were as follows:

"Some Observations on the Quality of the Radiographic Image" by *G. d'Herbemont*
"Behaviour of Ultrasonics in the presence of various types of possible weld defects" by *P. Bastien and M. Evrard*
"Radiographic Inspection Techniques for Weld-Examination" by *R. Halmshaw*
"Ultrasonic Inspection of Welding" by *D. O. Sproule*
"Destructive Testing for Quality Control in Weldments" by *R. E. Lismer*

Attendance at the meetings on Friday was smaller. In the morning the three following papers were presented and discussed, under the chairmanship of Monsieur Audige, a Past-President of the Société des Ingénieurs Soudeurs:

"Control of the Construction of Penstocks in the Shops and on Site" by *R. Bouchayer, G. Rambaud and J. Picard*
"Fusion Welding in General Engineering Works" by *B. K. Barber*
"The Control of Quality of Brazed Assemblies" by *F. Daintith*

At the end of this meeting, the President of the Institute thanked all the French visitors for their attendance and took leave of them, having to depart for the United States on the following day. Monsieur de Jessey replied, expressing the thanks and congratulations of the French party.

In the afternoon, Mr. H. Gardner presided over the closing session, when the three following papers were presented and discussed:

"Control of Spot Welding in the Aircraft Industry" by *H. Lefebvre*
"Spot Welding of Primary Aircraft Structures" by *N. K. Gardner*
"Standards of Welding for Airframe and Aircraft Engine Parts" by *F. J. Tranter*

The Secretary of the Institute, Mr. G. Parsloe, speaking in English and French, closed the Joint Meeting, offering the thanks and good wishes of the Institute to its visitors.

A full report of the discussion on these important papers will appear in a later issue of the *Journal*.

SCHOOL OF WELDING TECHNOLOGY

The following evening courses have been arranged for the end of January.

Detail Design of Welded Structural Steelwork: 19th January–22nd March (Institute of Welding, 54 Princes Gate, London, S.W.7.)

This course has been planned to provide those who are keen to keep abreast of technical progress with the specialized knowledge that is so important to the designer and detailer of welded structures. It will deal with the design, in detail, of welded members and joints from the view-point of technical efficiency, ease of fabrication and erection, and cost; it will not cover the theoretical analysis of the structure.

Design for Welding in Heavy Engineering: January–March commencing on 21st January (Weir Hall, Institution of Engineers and Shipbuilders in Scotland, 39 Elmbank Crescent, Glasgow, C.2.)

This is the first of a series of new evening Design Courses to be organized for designers and draughtsmen in the heavy engineering industry. Its purpose is to provide information from leading welding experts about all the factors on which the efficient design of a welded structure depends, so that one can derive maximum benefit from the technical and economic advantages offered by welded construction.

NEWS OF MEMBERS

Mr. **F. C. Braby**, M.C., D.L., M.I.Mech.E. (Chairman and Managing Director of Fredk. Braby & Co. Ltd.) has been elected Chairman of the Council of The British Non-Ferrous Metals Research Association to succeed Dr. Maurice Cook, C.B.E., who retires from office on 31st December.

Mr. **K. Holt** (Sales Manager) and Mr. **R. Gardner** (Works Manager) of Thos. Marshall & Son Ltd., Leeds, have been appointed to the Board of the Company, Mr. Gardner is Chairman of the Leeds and District Branch of the Institute.

Obituary

The Council regret to record the deaths of:

Mr. **B. R. Millar**, senior partner of Thomas Millar & Co. (Engineers) Ltd., Glasgow. (West of Scotland Branch, Member 1935.)

Mr. **Stanley Hirst**, Chairman and Managing Director of Henry Brook & Co. Ltd., Huddersfield, in October 1959. (Leeds Branch, Member 1941.)

BRANCH NEWS

East Midlands

The Branch held its Annual Dinner at the Victoria Station Hotel, Nottingham, on Monday, 19th October last, when about 100 members and guests assembled under the Chairmanship of Mr. K. H. Ewen. The toast of the Institute and its East Midlands Branch was proposed by Professor J. A. Pope in a speech which made a special plea for research in the fundamentals of welding. The President, Mr. E. Seymour-Semper, who responded, spoke mainly of the Institute's educational work and in particular the School of Welding Technology. The toast of the Guests was proposed by the Branch President, Mr. H. L. Palmer. In responding, Mr. H. L. Haslegrave, the Principal of Loughborough College, made some interesting comments on the progress of engineering education in Russia, which he had recently visited with a team of engineers.

East of Scotland

A joint meeting with the Institution of Structural Engineers was held on 3rd November with an audience of about 60. The meeting was presided over by Mr. J. F. Wheeldon, Branch Chairman, and Mr. C. N. Kington lectured on "*The 250 feet diameter radio telescope at Jodrell Bank*". This was followed by a film "*The Inquisitive Giant*" and Mr. Kington described some of the uses of the telescope in collection, magnification and concentration of radio waves and the tracking of satellites. The weight of the bowl and supports is 800 tons and the size and shape was determined by the requirements of the scientists. The total weight of the telescope is 2000 tons. For economic reasons, and because of the complexity of joints on site, the structure is mainly bolted or riveted, but welding was used for prefabricating parts of trunnions.

Questions asked related mainly to the balancing of the bowl, and the greater use of welding and of tubular construction. The maintenance and operating factor was said to be 22 hours per day, 7 days per week.

R.D.B.

Manchester

"*Fabrication of the Bradwell Heat Exchangers*" was the subject of the paper given by Mr. B. K. Twigg at the Third Meeting of the Branch held on Wednesday, 4th November, at the College of Science and Technology.

Members and visitors were taken through the many stages of manufacture of these large vessels, and the methods employed in resolving various problems relating to these stages were fully explained and illustrated by slides.

One very interesting feature of Mr. Twigg's paper was the description of the method adopted for transporting these fabrications from their place of manufacture to their final destination. This was effected by launching them into the sea.

Careful planning of production in relation to high tides for their successful launching was essential, and it is a tribute to the manufacturer's organization that each of the twelve exchangers supplied for this vast project was floated without a hitch for their long journey from Tyneside to the Thames.

The number and variety of questions put to Mr. Twigg at the conclusion of his paper, demonstrated the keen interest in welded fabrication by all those who attended.

A.C.W.

North London

On Wednesday, 21st October, Mr. J. Hinde, F.I.M., gave a lecture on "*The Welding of Nickel and High Nickel Alloys*". Methods of approach arose, covering preparation, various classes and types of alloys, and welding processes for their fabrication, including oxy-acetylene, argonarc, metal-arc, and in some applications, resistance welding. The lecture was illustrated by slides, showing types of fabrications possible; mainly chemical plant, pickling vats, and equipment, and fabrications applicable to corrosive acids. The question time proved very interesting, and the problem of stresses created by the welding processes was discussed. Heat treatment did not improve the weld qualities of N.75, and slag left on or adjacent to welds would cause corrosion. The weldability of Nimonic 105 brought a negative answer, owing to fundamental metallurgical problems; N.75 and 80 could be flash-butt welded. The vote of thanks for a most interesting and well delivered lecture was proposed by Mr. Coplestone.

On Wednesday, 18th November, the subject for the evening was "*Demonstrations and Applications of Stud Welding*", which was ably covered by Mr. R. W. Taylor and a team of demonstrators. The meeting was opened at the Institute, and members and visitors proceeded to the Imperial College of Science to see demonstrations of aluminium, copper, and steel used in the stud welding process. Up to 1 in. dia. studs were used, and the entire demonstration was very impressive. On their return to the Institute, members and visitors heard a lecture by Mr. Taylor, which was followed by question time. Slides were used to illustrate the many points covered, and the large audience present recorded their appreciation for an interesting and instructive evening.

R.B.W.

Preston

Branch Treasurership

Mr. P. Baron has been compelled through ill health to resign the office of Honorary Treasurer of the Branch, which he has held since it was started. The Branch Committee has received his resignation with great regret, and recorded its indebtedness to Mr. Baron, and its best wishes for his early recovery to normal health. Mr. Baron's successor is Mr. F. Westall.

South Western

In view of the popularity of the "Welding Forum" previously held, the Gloucestershire Engineering Society enjoyed a 'repeat' on 6th November at Gloucestershire Technical College at the invitation of the South Western Branch of the Institute.

The panel was under the chairmanship of Mr. T. M. B. Sessions (Messrs. Fielding and Platt Ltd.) and the panel consisted of: Mr. W. K. B. Marshall (Messrs. Rockweld Ltd.), Mr. L. B. Smith (British Oxygen Gases Ltd.), Mr. F. J. Wilkinson (Bristol Aero Jets Ltd.), and Mr. J. Hinde (Mond Nickel Co.).

The audience kept up a sequence of questions relating to the various problems and applications of welding and techniques employed in everyday engineering.

A vote of thanks was given by Mr. H. C. R. Pegler, President of the Gloucestershire Engineering Society, and in reply, the panel thanked the members of the audience for the variety and standard of questions put to them.

West of Scotland

A joint meeting of the West of Scotland Branch and the Institution of Structural Engineers' Scottish Branch, was held on Wednesday, 18th November 1959 in the offices of the Institution of Engineers and Shipbuilding, Glasgow.

A paper entitled "*Materials and Fabrication Problems in Reactor Technology*" was presented by A. Prince, to an audience of 87, under the chairmanship of D. B. Kimber.

The paper dealt with the work at present in progress at the Hunterston Nuclear Energy Generating Station in Ayrshire, and lantern slides showing shop preparation for the reactor plate-work, site welding and methods adopted for securing site tests on welded joints, revealed the tremendous job being undertaken in the construction work. The speaker emphasized that the problems associated with the fabrication of work of this kind were centred around the selection of suitable fabricating materials, such as the plate, forgings, and welding rods.

Great care had to be exercised in selecting the materials and a very stringent test applied in every case. The steel manufacturers had succeeded in supplying a plate material capable of meeting the standard required in the Charpy test—good behaviour at the transition temperature level, and freedom from brittle failure. Similar standards had been attained by the forgemasters, but the same could not be said about the electrode manufacturers. The minimum value set down for the test procedure was difficult to obtain with electrodes at present being manufactured.

Other tests mentioned included the Robertson test and the wide plate test. Four speakers took part in a discussion which followed, and questions raised aroused an exchange of views on the relative value of the tests outlined.

At the close, the Chairman of the Institution of Structural Engineers expressed the appreciation of his members for the invitation to take part in the meeting.

H.H.M.

BRITISH WELDING RESEARCH ASSOCIATION

Head of Design Section

Mr. Tibor Haas, Dipl.Ing., A.M.I.Mech.E., A.F.R.Ae.S., has been appointed Head of the Design Section in the newly formed Members' Service Department of the Association.

Mr. Haas, who was educated at Prague University, Faculty of Mechanical and Aeronautical Engineering, has had many years' experience in mechanical engineering, light weight engineering design and heavy engineering.

Following a period as an industrial consultant on contracts throughout Europe he joined the Bristol Aircraft

Company of Filton where, for the past six years, he has held the position of Head of the Fatigue Department.

Design Advisory Service

The appointment of Mr. Haas as Head of the Design Section is a prelude to the setting-up of a Design Advisory Service—a proposal briefly mentioned by Sir Alexander Fleck in his address to BWRA members at Open Day in July 1959, and outlined in the 14th Annual Report of the Association.

The formation of this Service has now been announced, and members of the Association can now receive guidance on welding design based on the wide knowledge and extensive experience of the Association's Officers.

Since the closing of the war-time Advisory Welding Service of the Ministry of Supply, there has been no central authority able to advise fabricators and designers on the essential principles that make sound design for welding. But, since the setting up of the Development and Information Section of the Association, there have been many consultations between the staff and members' designers to give guidance and advice on welding matters. Helpful as this service has been, it is but a very small part of what needs to be done in "education for welding".

All too often the capabilities and limitations of the various welding processes have not been fully appreciated; the "strength without beauty" of the older type of construction has frequently been replaced by the "beauty without strength" of the welded construction—many times with consequent failure and a condemnation of welding as a whole. Yet it is generally true that it is bad design that should be condemned and not the welding process. New concepts of design are needed for welded fabrication, and recognition and acceptance of this fact should lead to safer and more economic structures in the future.

The Association is now enabled very considerably to extend its services in this direction by a special grant from the Department of Scientific and Industrial Research. This is being provided with the condition that industry (through the membership of the Association) also contributes additional funds to the sum offered.

The Design Advisory Service will now be able to show BWRA members how research work can be applied at all stages of design, from the choice of the right materials and selection of the most suitable welding process to the correct method of assembly.

The behaviour of constructions under dynamic conditions of loading is one particular problem about which most designers are ignorant. To help provide factual information on possible loading conditions in service, the Design Advisory Service is to make available site investigating facilities by means of a mobile laboratory, which will be fitted with strain-gauge equipment and recording devices for measuring stresses and loads under practical conditions.

In addition, the equipment in the Association's Fatigue Testing Laboratory is being extended so that it can deal with actual tests on all the possible large-scale (and some times full size) welded components and structures.

It has been emphasized by the Director of the Association, Dr. Richard Weck, that the new service in no way displaces the specialized advice and guidance that already can be obtained from the recognized experienced welding consultants. It will, in fact, complement these existing facilities.

This new service to the welding industry, together with training facilities offered by the School of Welding Technology of the Institute of Welding, make a serious and important contribution to the progress of welded construction in Great Britain.

INTERNATIONAL INSTITUTE OF WELDING

1960 Annual Assembly

The Governing Council of the IIW has made the Belgian Institute of Welding responsible for the organization of the 1960 Annual Assembly, which will be held from **11th to 18th June** at Liège. It will be followed from 19th to 26th June, by industrial and scientific visits.

No public session will be included in the programme for the Assembly, but four of the technical commissions are holding colloquia, at which a number of communications will be presented by rapporteur and discussed.

Metal spraying is one of the subjects (Commission I – Gas welding and Allied Processes), and others are the inspection and testing of welds on plastics (Commission V – Testing, Measurement and Control of Welds) and the influence of residual stresses on the stability of welded structures (Commission X – Residual Stresses and Stress Relieving).

At the fourth colloquium the Koudremont lecture, delivered annually to Commission IX – The Behaviour of Metals subjected to Welding, will be discussed. This year the lecturer will be Professor Soete (Belgium).

As at preceding assemblies, arrangements will be made to show films on welding subjects, and there will be an exhibition of publications and of photographs of welded structures. Because of the exceptionally favourable possibilities offered by the Liège Palais des Congrès, where all the meetings in connection with the Assembly will take place, it will be possible to give particular emphasis to these items. The Belgian Institute of Welding hopes that member societies will assist by giving consideration now to the collection of new films, and by ensuring that the exhibitions of publications and photographs will be as successful as possible.

A special programme will be arranged for ladies and others accompanying delegates; events in this programme will take place in the town of Liège, and in the beautiful surrounding countryside of the Ardennes.

List of terms used in ultrasonic testing

Commission V "Testing, Measurement and Control of Welds" of the IIW has now published a "*List of terms used in ultrasonic testing*". This 28 page booklet, which contains the same terms in English, French, and German, is divided into two sections:

(1) A list of the terms in the three languages, arranged in logical order. Each term is numbered from 9000 to 9999, the numbers being the same for each language. The list is divided into chapters to assist in the finding of a given term. It is followed by some pages of figures illustrating the meaning of certain terms.

(2) An alphabetical classification for each language, in which each term is followed by the number allocated to the same term in the logical list.

Copies of the booklet, price 7*s* inclusive of postage, are obtainable from The Honorary Secretary, Joint Committee for International Welding Relations, 54 Princes Gate, London S.W.7.

NEW PLANT AND EQUIPMENT

Bronze welding rod

An all-purpose, flux-coated bronze welding rod, the LECO No. 1, for oxy-acetylene welding and brazing, is now being marketed by the Lincoln Electric Co. Ltd.

The wire is a high quality grade, degasified and deoxidized, and has a durable flux coating. It melts rapidly and tins easily, and has free flowing properties, although it solidifies quickly enough for the deposit to be controlled.

This welding rod has been designed for application to the braze welding of steel, cast iron and malleable iron, and other higher melting point metals; for the fusion welding of brass, bronze, and copper alloys; and for building-up of worn surfaces.

New Welding Transformer

The replacement of copper wire in the windings by specially treated aluminium

foil has given the new Balarc 175 welding transformer an unusually high power/weight ratio.

The unit, which is made by the Ashton-Young Engineering Development Co. of Kingston-on-Thames, is rated at 175 amp for continuous operation, and yet it weighs only 67 lb. Electrolytic treatment of the aluminium foil allows the temperature to rise safely to about 500°C.

The Balarc 175 meets the specifications of many countries, including the U.K., U.S.A., and Germany, and is suitable for continuous operation in high ambient temperatures and under adverse climatic conditions. Every unit is chemically treated to prevent the ingress of moisture and dirt.

A 2-pole, 6-way roller switch controls the current at both 80 V and 50 V open circuit and with an arc voltage of 23 V. The unit draws a maximum of 40 amp from 190–250 V a.c. mains, single-phase, at the normal 50 cycles.

Electrode holders

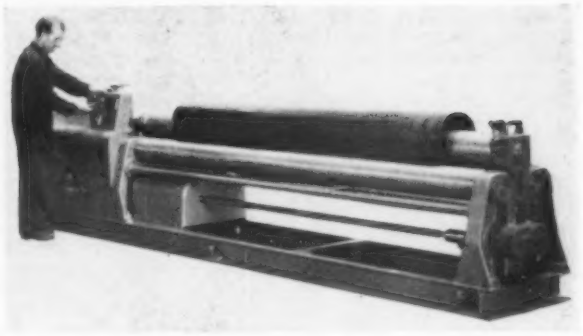
Quasi-Arc Ltd. of Bilston have introduced two electrode holders for manual welding, based on a study of welders' requirements and produced after extensive field trials. A feature of the holders is the new lever shape, which is angled close to the handle to give better access to confined spaces. The handle and lever both have D-shaped cross-sections providing a compact, well-balanced grip for welding in any position. The holders are designed so that they can be held with a long grip at the end of the handle when using high currents, or with a short grip close to the jaws to give a good control of the electrode tip. The holders keep exceptionally cool even after long periods of welding. Weight of the "Handicool" 600 is 21 ounces, and of the "Handicool" 600 (Insulated) 23 ounces. Maximum current is 600 amps, and the range of electrode sizes extends from 10 s.w.g. to $\frac{1}{4}$ in.

Contact welding electrode

Philips Electrical Ltd. are now marketing a new zircon iron-powder coated electrode which has been specially developed for rapid trouble-free contact welding of all mild and low-alloy steels in downhand, standing fillet, and horizontal-vertical positions.

The mechanical properties, X-ray qualities, and the extremely high impact values, which are in excess of conventional low-hydrogen electrodes, obtained with the new C.16 electrode, make it particularly suitable for the demanding conditions of nuclear engineering. It has been approved by the Ministry of Transport and Lloyd's Register of Shipping.

The deposit is flat or slightly convex, and the slag is self-detaching.



Edwards TRP plate bending rolls

Quick starting for argon tungsten-arc welding

The use of high-frequency current has generally been the accepted method of arc starting for argonarc spot welding. A new development of Union Carbide, 'Pilot-Arc' starting, has now made it possible to get positive arc starting in a split second whenever required. This is of particular importance in mechanized installations where torches must be fired according to a predetermined schedule.

With this device a ball of incandescent gas is maintained within the cup so that the arc is always ready to fire as soon as potential is applied between torch and workpiece. The 'Pilot-Arc' current is adjusted to about 5 amp, and is maintained between electrode and water-cooled nozzle. A small quantity of shielding gas flows continuously and protects the electrode while the pilot is on.

When the torch is brought into welding position against the work and the trigger is pulled, the Pilot-Arc provides a path for the main welding current which instantaneously flashes across to the work and makes the spot weld.

Gas regulators

A new range of single-stage oxygen and acetylene regulators, S.O.R.I. and S.A.R.I. has been produced by British Oxygen Gases Ltd. for general cutting and on-site work. The pressure gauges are of the small dial type with a range of 0–3000 lb/sq.in. oxygen and 0–600 lb/sq.in. acetylene. A feature of the regulators is that the joints are dry.

Bending rolls

A range of pyramid type plate bending rolls, Model TRP, with capacities of 6, 8 and 10 ft by $\frac{3}{8}$, $\frac{1}{4}$ and $\frac{3}{16}$ in. mild steel respectively, is now being produced by F. J. Edwards Ltd. at their Chard Works. Power control covers the operations of adjustment to the top pressure roller for various radii, and the actual feeding of the plates through the rollers. A screw-operated rise and fall self-support to the top pressure roller, around which plates are formed, frees the plate for withdrawal from the machine. The two bottom rollers are supported in fixed bearings and at close centres, so that the flats at start and finish of the work are reduced to a minimum.

Limit switches with adjustable stops give very precise control to the top roller. Reverse direction of feed is obtained by a joystick reversing controller on the main control panel. A safety device prevents the machine from running unless the top pressure roller support is released.

Suspension device

A plaited nylon 'suspensor' made by Worcester Fabrications Co. may have useful applications for the temporary or permanent support of cables and small pipelines.

The device, known as the G.E.M. Suspensor, can withstand a load of 650 lb, and the nature of the weave is such that once in place it cannot slip, though it can be released with ease when the load is removed. There are three different ends to provide a choice of anchoring methods.

STUB ENDS

▶The use of electricity for preheating and stress relieving is featured in an article in No. 53 issue of "*Wiggin Nickel Alloys*".

▶The Smith's Centre, a new central showroom at the headquarters building of S. Smith & Sons (England) Ltd., Cricklewood, was recently opened by H.R.H. The Duke of Edinburgh. The centre is designed to show a selected range of products from every side of the business, grouped in their specific industries or applications.

▶The Public Works Department of Brunei, North Borneo, have installed Quasi-Arc submerged-arc equipment for the repair of worn tractor links, idlers and rollers.

▶The 42,000 ton d.w. oil tanker *British Power* built by Harland and Wolff Ltd. for the BP Tanker Co. Ltd., has recently completed sea trials and has been handed over to the Owners. She is a sister to the *British Statesman* completed earlier in the year.

▶Mr. C. W. V. Davis, a Director of Research and Control Instruments Ltd., will head the newly-formed Industrial Products Division of that Company, which will in future distribute Philips arc welding supplies, filtration equipment and high frequency generators.

▶Mr. Michael J. S. Clapham has been appointed Chairman of I.C.I. Metals Division from 1st January, 1960, in succession to Dr. Maurice Cook, C.B.E., who retires from the Company's service on 31st December, 1959.

▶The British Productivity Council have published for The National Union of Manufacturers Advisory Service a booklet describing how the Advisory Service has assisted smaller firms, ranging in size from 6 to 300 employees, to achieve substantial increases in productivity and reduction in manufacturing costs.

▶Dr. A. B. Kinzel, Vice-President of Research of the Union Carbide Corporation is the recipient of the American Industrial Institute Medal for 1960. The medal has been awarded annually since 1945 to honour "outstanding accomplishment in leadership in or management of industrial research which contributes broadly to the development of industry or the public welfare." The presentation of the medal to Dr. Kinzel will take place next May at the annual meeting of the Industrial Research Institute.

CORRECTION

Metal Spraying

In the November 1959 issue (p. 550) it was stated that Mr. Ballard's paper was presented to the Netherlands Welding Society. It was in fact contributed to the Advisory Office for Gas Welding and Allied Processes (Adviesbureau Voor Autogene Lastechniek).

Mr. Eric Courtney

By the omission of the word "late" after Mr. Courtney's name in the note on a New Anglo-Dutch firm (Nov. 1959 issue, p. 552) it was made to appear that he was still a Director of Courtburn Positioners Ltd. In fact, Mr. Courtney severed all connections with that company on 31st May 1959.

DIARY

4th Jan.—Sheffield—Annual Dinner (Royal Victoria Hotel)

4th–15th Jan.—School of Welding Technology—Course D.27 "Practical ultrasonics for operators".

5th–7th Jan.—Institution of Mechanical Engineers—Symposium on recent mechanical engineering developments in automatic control.

5th Jan.—Institute of Welding—Christmas Lecture to Young People—"*The Magic Arc*" by Professor H. O'Neill (54 Princes Gate, 11.0 a.m.)

6th Jan.—Manchester—"*Inspection and interpretation of welding codes*" by C. Atkinson (College of Technology, 7.15 p.m.)

7th Jan.—North Eastern (Tyneside)—"*Welding of the boilers for Bradwell atomic power station*" by P. E. M. Jones and K. Twigg (Mining Institute, Neville Hall, Newcastle, 7.0 p.m.)

10th Jan.—East Midlands—Paper on welding in BTC diesel development programme by G. Walton (Loughborough College, 7.15 p.m.)

11th Jan.—West Wales—"*Welding and the draughtsman*" by H. B. Merriman (Steel Company of Wales, Old General Office, Port Talbot, 7.0 p.m.)
Sheffield—"*High vacuum: its production and use*" by L. G. Beckett (Joint meeting with the Rotherham Engineering Society) (Rotherham Technical College, 7.15 p.m.)

12th Jan.—Eastern Counties—Two 16 mm films—Foundry Practice Craftsmen welders, J. Strevens (for questions) (Eastern Electricity Board, Ipswich).
East Wales—"*Welding and the draughtsman*" by H. B. Merriman (Education and Welfare Dept. of Richard Thomas & Baldwin Ltd., Ebbw Vale, 7.0 p.m.)
Liverpool—"*Welding and physical metallurgy*" by T. Fullwood (College of Technology, 7.30 p.m.)
South Western—"*The development and welding of admiralty steels*" by Mr. Gundray (Stothert & Pitt Ltd., Bath, 7.15 p.m.)

13th Jan.—South London—"*High temperature brazing—A modern joining method*" by E. R. Perry (54 Princes Gate, 7.30 p.m.)

14th Jan.—Preston—"*Oxy-acetylene welding of non-ferrous metals*" by E. A. Foreman (Municipal College, Burnley, 7.15 p.m.)

20th Jan.—North Eastern (Tees-side)—Open discussion (Cleveland Scientific and Technical Institution, Corporation Road, Middlesborough, 7.30 p.m.)
North London—"*Further developments in the welding of atomic plant*" by A. Prince (54 Princes Gate, 7.30 p.m.)
West of Scotland—"*Photo-electric automatic flame-profiling*" by J. Marsh (Institute of Engineers and Shipbuilders, Glasgow, C.2, 7.0 p.m.)

21st Jan.—Leeds—"*Fabrication of the Bradwell heat exchangers*" by B. K. Twigg and A. R. Muir (Great Northern Hotel, 7.30 p.m.)
Southern Counties—"*Welding of spike fuel elements*" by E. G. King (Technical College, St. Mary's, Southampton)

22nd Jan.—Birmingham—"*Improved steels and their weldability*" by F. J. Wilkinson (Grand Hotel, 7.30 p.m.)
Wolverhampton—"*Some aspects of electronic control in welding*" by J. Teece (Wulfrunians Club, 7.30 p.m.)

25th–29th Jan.—School of Welding Technology—Course D.4/1 "Resistance Welding"

26th Jan.—East of Scotland—"*Non-destructive testing of welds*" by G. Blyth (Brown Bros. & Co. Ltd., Rosebank Ironworks, Edinburgh, 7.30 p.m.)

27th Jan.—South London (Medway Section)—Demonstration of new welding processes (Medway College of Technology, Chatham, 2.0–4.30 p.m. and 6.30–9.0 p.m.)

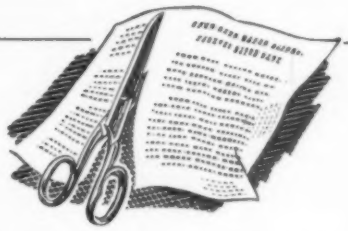
2nd Feb.—East Midlands—"*Welding and the draughtsman*" by H. B. Merriman (Victoria Station Hotel, Nottingham, 7.15 p.m.)
Eastern Counties—"*Welding in the Structural Industry*" by F. Brooksbank (City College of Art School, Norwich)

3rd Feb.—Manchester—"*Welding in the modernization of the railways*" by P. S. A. Ferridge and G. D. S. Alley (College of Technology, 7.15 p.m.)

4th Feb.—North Eastern (Tyneside)—"*Welding and ship salvage*" by E. P. Wilson (Mining Institute, Neville Hall, Newcastle, 7.0 p.m.)
Institute of Welding—Conversazione "IIW work in hand" (54 Princes Gate, 4.0 p.m.)

8th Feb.—*West Wales*—"*Modernization of British Railways (Western Region)*" by M. G. R. Smith and R. A. Smeddle (Swansea Technical College, 7.0 p.m.)
Sheffield—"*A practical approach to the problem of distortion*" by J. S. Waring (Grand Hotel, 7.15 p.m.)

Current WELDING LITERATURE



Book Reviews
Additions to the Institute Library
Trade Publications

CONTENTS OF PERIODICALS RECEIVED

This section is intended to give a survey of the current welding literature received by the Institute of Welding Library. The contents lists are not exhaustive; only the main articles in welding journals are listed, and reprints from other journals and short notes are generally excluded. In addition, welding articles from other periodicals are listed. Abstracts of welding literature are given in the Bibliographical Bulletin of Welding and Allied Processes, *published by the International Institute of Welding, and details of this may be obtained from the Secretary of the Institute of Welding.*

Welding Journals

Australian Welding Journal, 1959, vol. 2, No. 10, June

Static brittle-fracture initiation yield at net stress 40% of yield, C. Mylonas, D. C. Drucker, and J. D. Brunton (26–29)

Canadian Welder, 1959, vol. 50, August

A short history of welding, A. A. Thompson (6–7)

1959 Directory—Product classification, Product listing, Manufacturer-distributor sections (11–48)

Canadian Welder, 1959, vol. 50, September

Automatic flame profiling, W. H. Pearse (10–12)

'Enclosed' welding: Can this concept be widely applied? J. A. N. Clevers and G. Zoethout (14–15)

Spray welding reclaims diesel engine valves (16–17)

Japan Welding Society Journal, 1959, vol. 28, July

Theory and review of recent studies concerning unconsumable arc welding electrodes, part 3, A. Uchida

Automatic submerged arc welding of steel sheets for railway cars, M. Yabe and others

How to avoid cracking in welded cast iron, part 1, I. Onishi and M. Takenmasu

Study on welding of low alloy high strength steels, part 3, Y. Ito and others

Characteristics of consumptions of iron anode and cathode of arc in nitrogen and chlorine, M. Ozawa and others

On the porosity of weld metal by gas-shielded metal-arc welding, part 1, H. Sekiguchi

Submerged arc welding of steel, T. Kobayashi and H. Okajima

Welding of plastics, part 2, I. Onishi and H. Kimura

Przeglad Spawalnictwa (Poland), 1959, September

Weldability and application of Al-Mg alloys, Z. Frackowiak

Czechoslovakian welding equipment, P. Fucik and E. A. Juffy

Schweisstechnik (Germany), 1959, vol. 9, September

The behaviour of all-welded framework under dynamic load, A. Neumann (328–335)

Effect of moisture in CO_2 on the CO_2 arc welding process, G. Hebold (336–337)

Resistance welding in the industrial production, G. Deubel (338–343)

The application of flash butt welding in the manufacture of feed-water preheaters for heavy-duty boilers, G. Bauch (343–345)

Vertical up welding technique and examples of application for gray cast-iron welding with pre- and postheating, G. Morawietz (346).

Soudage et Techniques Connexes (France), 1959, vol. 13 July–August

Workshop and site control of penstock construction, R. Bouchayar, G. Rambaud and J. Picard (249–259)

Control of spot welding in the aeronautical industry, H. Lefebvre (261–274)

Example of the organization of a control service for the construction of welded ships, boilers and frames, P. Pouchet and others (277–284)

Some observations with regard to the quality of the X-ray images, G. D'Herbemont

Improvement of the weldability of aluminium-magnesium alloys, R. Chevigny

Quality control of welded construction in the field of piping for the transport and distribution of petroleum products and combustible gases, H. Gerbeaux

La Soudure Electrique (Belgium), 1959. No. 21

Metallurgical weldability tests, J. Cabelka (3–11)

The scope of resistance welding; part III, Experiments in spot welding, J. Belotte and E. Lheureux (13–21)

Multispot welding machines, G. E. Lheureux (23–32)

Welding and Metal Fabrication, 1959, vol. 27, August–September

Progress at Hunterston (300–306)

A high speed all-welded steel passenger launch, R. Du Cane (307–315)

Industrial uses of electroslag welding in the Soviet Union, E. Bishop (316–319)

Truck cab fabrication, T. J. Palmer (320–324)

Development of a transition weld between ferritic and austenitic superheater tubing, B. Löfbald and G. Lindh (325–330, 349)

Joining dissimilar metals, part 2, J. G. Young and A. A. Smith (331–339)

B.W.R.A. open day (346)

The dip transfer process (347–8)

Welding and Metal Fabrication, 1959, vol. 27, October

The production of earthmoving equipment: A report on the new Caterpillar Works (354–361)

Aluminium fabrication in rail transport, N. T. Burgess (362–369)

Fabricating steel for Atlas missiles (370–372)

The Hinkley Point project (373–378)

Automatic spot welding on the new B.M.C. baby cars; details of some specially designed spot welding machines (379–384)

Dome end fabrication: details of special cutting equipment used at Bradwell (385–386)

Welding Design and Fabrication (U.S.A.), 1959, vol. 32 September

Techniques for resistance welding tiny, dissimilar metal parts (44–45)

Will brazed molybdenum replace the "superalloys" in jet engine parts? G. S. Hoppin (46–48, 78–79)

R.W.A.A. proposes a new numbering system for resistance welding electrodes (49)

Welding platens can stop distortion (50, 80–81)

Careful stress relieving is the key to quality welding of alloy piping (53, 78)

A design approach to quality control of structural welding, C. I. Orr (54–55, 82)

Flash butt welding cuts jet ring costs 65%, H. J. Novak (56–57)

Know your filler metals, O. T. Barnett (60–61)

The strength of arc welded joints (62–65)

Plating speeds aluminium soldering, W. J. Ross (66)

Welding Engineer (U.S.A.), 1959, vol. 44, September

How to get better stainless steel welds, W. L. Wilcox (32–34)

Spot welding procedures and design, part 1, L. F. Spencer (35–39)

Understanding weld distortion and shrinkage, G. W. Gardner (40–42)

Tig welding, aluminium used to manufacture durable and economical transformer cases (43)

Modern geodesic dome replaces an old-fashioned railcar repair plant, (44–45)

Importance of considering welding stresses reviewed, R. E. Holt (48–49)

Welding helps to launch the first nuclear-powered merchant ship, N. S. Savannah (50–51)

Hand welding is still the most economical in many cases (52–53)

Welded-steel flat cars increase piggyback efficiency, G. C. Shields (84)

Welded pipe to avert water shortage in India, W. A. Heath (90–91)

Welding Journal (U.S.A.), 1959, vol. 38, September

Maintaining superior weld quality in a plutonium production plant, C. D. Brons (853–859)

Welded steel transfer bridge for loading railroad-car floats, M. L. Koehler (860–864)

Unitized body construction, K. M. Sims (865–869)

Plasma- a substitute for the oxy-fuel flame, J. A. Browning (870–875)

A portable pipe-welding machine, H. L. Saunders (876–884)

Facelifting . . . freighter converted to luxury liner, S. Richardson (885–886)

Economies effected in the semi automatic welding of piling, J. F. Gulley (887–888)

Feasibility of titanium for welded missiles, C. E. Hartbower and others (345s–353s)

Spontaneously induced brittle fractures in welded-steel disks, A. Vinckier (354s–359s)

Spot welding of a 12%-Cr Martensitic stainless steel, E. F. Nippes and others (360s–366s)

Micromechanism of brittle fracture in a low-carbon steel, G. T. Hahn and others (367s–376s)

Development, properties and usability of low-hydrogen electrodes, D. C. Smith (377s–392s)

Welding News (Australia), 1959, Folio 99, July

Weld costing tables (11–14)

Having fuel gas problems? (15–18)

Crayfish, dollars and argonaut welding (20–21, 22)

Copper tubes for building industry save time and money (23)

Welding News (The Netherlands), 1959, No. 102, June

Philips electrodes obtain their approvals readily, M. D. Huisman

The various versions of the CO_2 automatic welding machine, J. Camerlink

Principles of the non destructive testing of materials

Welding Production (U.S.S.R. (translated by B.W.R.A.), 1959, No. 9, September

Arc plasma jet as a source of heat during material treatment, I. D. Kulagin and A. V. Nikolaev (1–4)

About heating by plasma in welding processes, K. V. Vasilyev and A. A. Isachenko (5–6)

Research weldability of heat resistant 12% steels, M. K. Shorshorov and others (6–10)

Research into arc welding of type ЭИ572 heat-resistant austenitic steel, V. A. Toropov (11–15)

About effect of ferrite phase on behaviour of weld and parent metal of welded joints made of chromium-nickel austenitic steels, A. E. Runov and K. V. Lyubavsky (15–19)

Impulse-butt welding of type 1X18H9 steel of small cross sections, D. M. Shashin and N. J. Palchuk (19–21)

Strength of structural members from type D16AT and B95AT spot-welded alloys under multi-static load, N. K. Andreev (21–22)

Wear-resisting welding up with lying plate electrode, B. M. Kontorov and N. V. Popova (27–30)

Supersonic welding of plastics, N. A. Olshansky and A. V. Mordvintseva (30–33)

Type ТЭ-10 diesel locomotive uniframe body welded, L. D. Veretnik and B. E. Chekulaev

Arc welding of studs to alloy steels, cast iron and non-ferrous metals, D. I. Vainboim (35–37)

Arc welding of turbine rotor of drying fan, J. I. Vodyanitsky and L. Z. Dolgitser (37–38)

Zvaranie (Czechoslovakia), 1959, vol. 8, September

Contribution to the stage analysis of welded joints Austenit-Ferrit, I. Hrivnak (258–261)

New fluxes VUS, St Horvath and E. Pikna (261–269)

Type 10 523 steel welding and employment in structural steelwork, (270–276)

Argon shielded arc welding, J. Novotny (276–282)

Welding press VUS 1000 (283)

Days of new techniques in CO_2 welding, L. Lanyi (285)

Other Journals

Electric arc welding; summary of proceedings at Birmingham University conference, 25 June 1959 (*Metal Industry*, 1959, vol. 95, 18 September, pp. 120–121)

How to repair grooved press rolls cheaply: Use of automatic Sigma welding in American press rolls (*Wiggin Nickel Alloys*, 1959, No. 53, pp. 22–23)

HF resistance welder achieves ultra-speeds (*Metalworking Production*, 1959, vol. 103, 2 October, pp. 1559–1560)

Spot welding the new Austin baby car (*Engineering*, 1959, vol. 188, 9 October, p. 330)

Heliarc offers better way to cut stainless steels (*Steel* (U.S.A.), 1959, vol. 145, 7 September, p. 93)

An introduction to silver brazing (*African Oxygen*, 1959, vol. 5, September, pp. 15–18)

Brazing aluminium alloys, Part I—Principles (*Metal Industry*, 1959, vol. 95, 25 September, pp. 135–8)

Brazing aluminium alloys, Part II—Practice (*Metal Industry*, 1959, vol. 95, 2 October, pp. 157–159, 166)

Honeycomb panels get braze in electric blanket set up, R. E. Crump (*Iron Age* (U.S.A.), 1959, vol. 184, 17 September, pp. 112–113)

The managing director and non-destructive testing, W.E.S. (*Solus Schall News on Non-Destructive Testing*, 1959, No. 14, September, pp. 3–7)

Polyethylene glycol as a vehicle for acid fluxes, W. R. Lewis (*Tin and its Uses*, 1959, No. 47, Summer, pp. 3–5)

The ABC of welding 'T-1' steel (*Steel* (U.S.A.), 1959, vol. 145, 14 September, pp. 164–5)

Fabrication techniques for tantalum, B. Payne (*Metalworking Production*, 1959, vol. 103, 25 September, pp. 1514–6)

Capacitor discharge process welds dissimilar metals (*Iron Age* (U.S.A.), 1959, vol. 184, 1 October, pp. 78–79)

Progress in the fusion welding of non-ferrous alloys, J. G. Young (*Sheet Metal Industries*, vol. 36, No. 388–9, August-September 1959, pp. 557–566)

Achievements to date in light-gauge building construction, F. W. Heathcote (*Sheet Metal Industries*, vol. 36, No. 388–9, August-September, pp. 595–600)

Thrust reversal for jet aircraft, R. H. Colley (*Journal of the Royal Aeronautical Society*, 1959, vol. 63, October, pp. 563–571)

Experimental investigation into the stress distribution in a band-reinforced pressure vessel, K. G. Mantle, N. Marshall and P. J. Palmer (*Proceedings of the Institution of Mechanical Engineers*, 1959, vol. 173, No. 4, pp. 123–133, Discussion pp. 134–141, Communications and Author's reply pp. 142–146)

First Berkeley vessel proved (*Nuclear Engineering*, 1959, vol. 4, October, pp. 362–3)

Welding a heavy duty trailer (*African Oxygen*, 1959, vol. 5, September, pp. 19–20)

Welding in Africa's largest car factory (*African Oxygen* (South Africa), 1959, vol. 5, September, pp. 21–23)

Bridge building in Ghana (*African Oxygen* (South Africa), 1959, vol. 5, September, pp. 26–27)

Welding problems in the chemical industry, A. Mauchan (*Engineering and Foundryman* (South Africa), 1959, vol. 25, No. 3, July, pp. 38–44)

'Weltexa' piping systems (*Nickel Bulletin*, 1959, No. 11, p. 23)

Cellular core panel structure upgrades steel doors, W. M. Leeser (*Iron Age* (U.S.A.), 1959, vol. 184, 3 September, pp. 80–82)

Scots nuclear power station uses heavy welds, huge crane (*Canadian Machinery*, 1959, vol. 70, September, pp. 133–134)

Welding and steel structures, A. J. P. Tucker (*Engineer and Foundryman* (South Africa), 1959, vol. 25, August, pp. 68–76)

Application of tubes to structures in South Africa, C. P. Sayce (*Engineer and Foundryman* (South Africa), 1959, vol. 25, August, pp. 79–84)

Stress in practical design, E. J. Dent (*Machinery Lloyd* (European edition), 1959, vol. 31, 3 October, pp. 32–33)

Producing thicker welds in a single run (*Engineering*, 1959, vol. 188, 2 October, pp. 298–299)

Welding ultra-thin metal parts, H. W. Ulmer (*Machine Shop Magazine*, 1959, vol. 20, September, pp. 519–521)

How manufacturing methods cope with future space projects (*Iron Age* (U.S.A.), 1959, vol. 184, 1 October, pp. 86–88)

Adhesive bonding, J. C. Merriam (*Materials in Design Engineering* (U.S.A.), 1959, vol. 50, September, pp. 113–128)

Bonded joints build new turbo-jet airframe, D. E. Winter (*Metalworking Production*, 1959, vol. 103, 25 September, pp. 1517–20)

Ultrasonic weld testing made easier, R. Hornung (*Engineering*, vol. 188, No. 4872, 4 September, pp. 141–2)

Hard surfacing strikes back (*Iron Age* (U.S.A.), 1959, vol. 184, 3 September, p. 83)

Stud weld for fast mounting (*Iron Age* (U.S.A.), 1959, vol. 184, 10 September, p. 180)

New tool for stress analysis (*Iron Age* (U.S.A.), 1959, vol. 184, 10 September, p. 192)

BOOKS AND PAMPHLETS

AMERICAN PETROLEUM INSTITUTE: Std 5A *Specification for casing, tubing, and drill pipe*. 22nd ed. New York, A.P.I., March 1958. (Price $1.00)
Std 5L *Specification for line pipe*. 17th ed. New York, A.P.I., March 1958. (Price $1.00)
Std 5LX *Specification for high-test line pipe*. 8th ed. New York, A.P.I., March 1958. (Price 50 cents)
Std 6D *Specification for steel gate, plug, and check valves for pipeline service*. 8th ed. New York, A.P.I., November 1958. (Price 50 cents)
Std 12D *Specification for large welded production tanks*. 7th ed. New York, A.P.I., August 1957. (Price 50 cents)
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RP 12H *Recommended practice for installation of new bottoms in old storage tanks*. New York, A.P.I., October 1957. (Price 50 cents)
RP 510 *Recommended practice for inspection, repair, and rating of unfired pressure vessels in service in petroleum refineries*. New York, A.P.I., July 1958. (Price $1.00)
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UNITED KINGDOM ATOMIC ENERGY AUTHORITY (Industrial Group): *Brazing graphite to Nilo-K tubes*, by B. Stapleton. Risley, I.G. HQ, 1959. (Price 2*s* 6*d*)

BRITISH STANDARDS INSTITUTION. BS.2614:1959. *Aluminium alloy sections for marine purposes*. London, B.S.I., 1959. (Price 12*s* 6*d*)

Requirements for two new series of aluminium alloy structural section for marine use are included in this revised edition of BS.2614:1955. They relate to plain tee-bars for welding and plain bulb plates for welding. Metric equivalents of all dimensions and tolerances have been calculated, and are included, for information, as separate tables.

GREAT BRITAIN. Ministry of Education. Science Museum. *Handlist of short titles of current periodicals in the Science Library*. 8th ed. London, H.M.S.O., 1957. (Price 15*s*)

GREAT BRITAIN. Ministry of Education. Science Museum. *Books on engineering: a subject catalogue of books in the Science Library*. London, H.M.S.O., 1957. (Price 16*s*)

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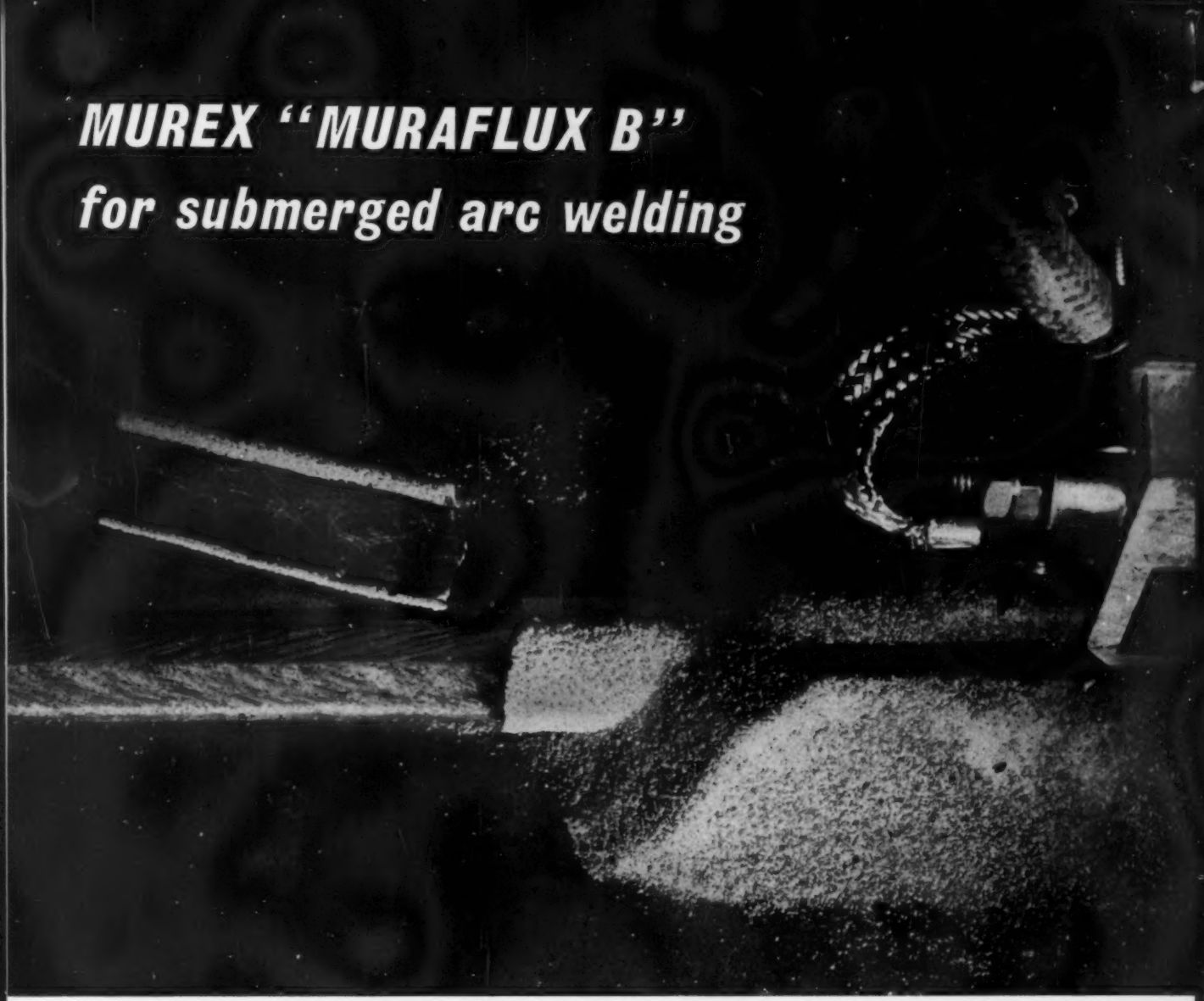

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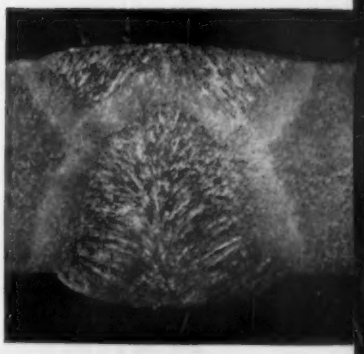
(Right) Photomacrograph of 1¼ inch thick steel plate butt welded in two passes with "Muraflux B" and ⅛ inch "Murawire W1", at 1,450 amp.



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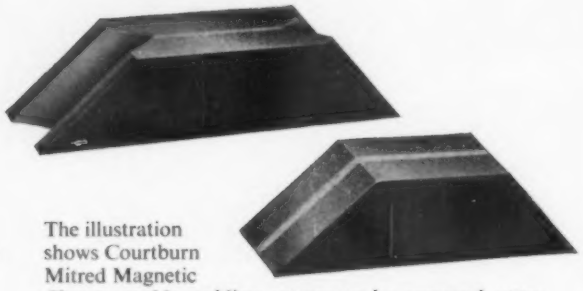

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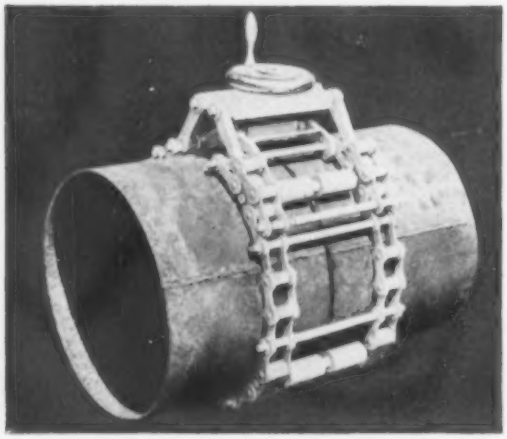

The illustration shows a model 612—Courtburn Adjustable Pipe Clamp. The clamp is holding and aligning two short lengths of 10 in. N.B. pipe. Ample accessibility is provided for tack welding all round the pipes, and space can be left between the pipe ends for the weld filler as required.

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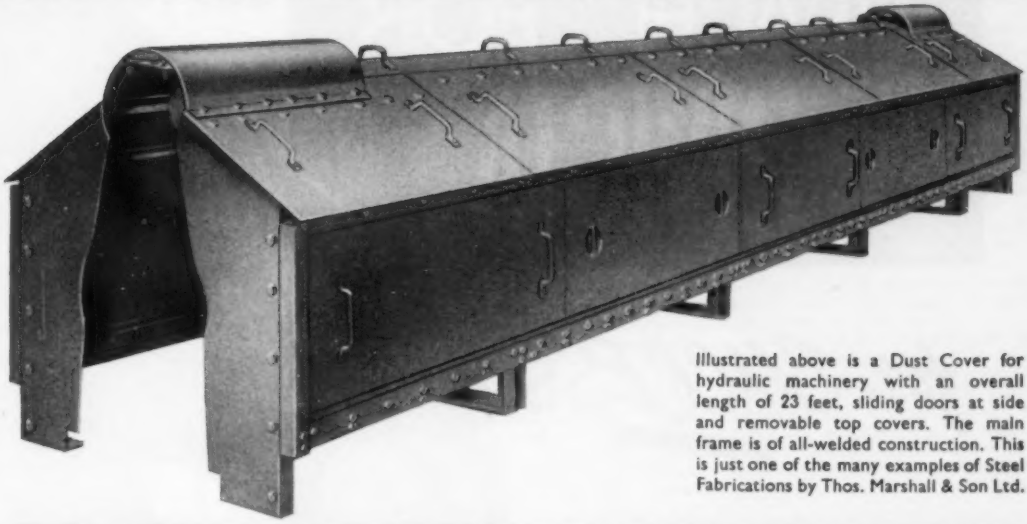

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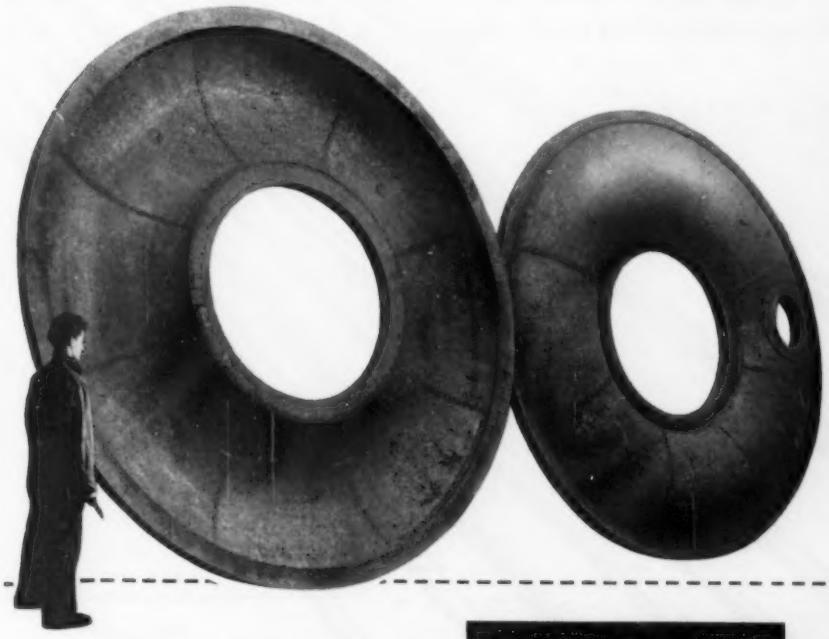

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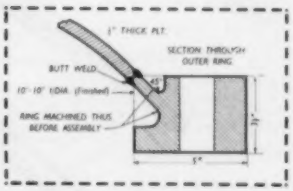

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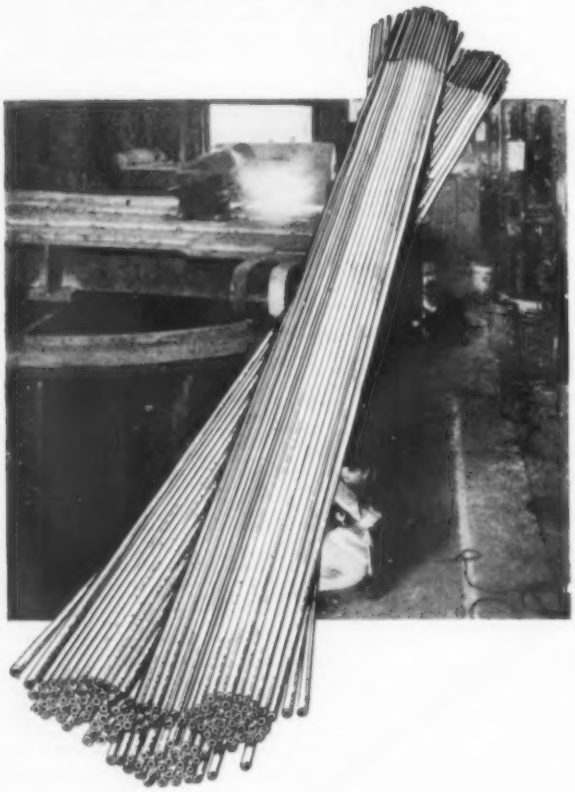

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ADVERTISERS' INDEX

Associated Electrical Industries:
Electronic Apparatus Division . . 14
Transformer Heating & Welding Division 8, 9
Booth, John, & Sons Ltd. 29
British Industrial Gases Ltd. . Inside front cover
British Oxygen Gases Ltd. Outside back cover 13
Classified Advertisements 32
Courtburn Positioners Ltd. . . . 28
Crompton-Parkinson 20
Distillers Co. Ltd., The 10
English Electric Co. Ltd., The . . . 22
Eutectic Welding Alloys Co. Ltd. . . 31
Head Wrightson Teesdale Ltd. . . . 19
Invicta Electrodes, Ltd. 30
Jenkins, Robert, & Co. Ltd. . . . 17
Johnson, Matthey & Co. Ltd. . . . 18
Johnson, Richard, & Nephew, Ltd. . . 32
Kodak Ltd. 5
Marconi Instruments Ltd. 16
Marshall, Thomas, & Son, Ltd. . . 29
Millspaugh Ltd. 15
Morris, B. O., Ltd. 4
Murex Welding Processes, Ltd. . . 27
Norton Grinding Wheel Co. Ltd. . . 24
Oxhycarbon Co. 26
Philips Electrical, Ltd. . . . . Inside back cover
Quasi-Arc Ltd. 2, 3
Rockweld, Ltd. 1
Saturn Industrial Gases Ltd. . . . 6
Siemens-Schuckert (Great Britain) Ltd. . 11
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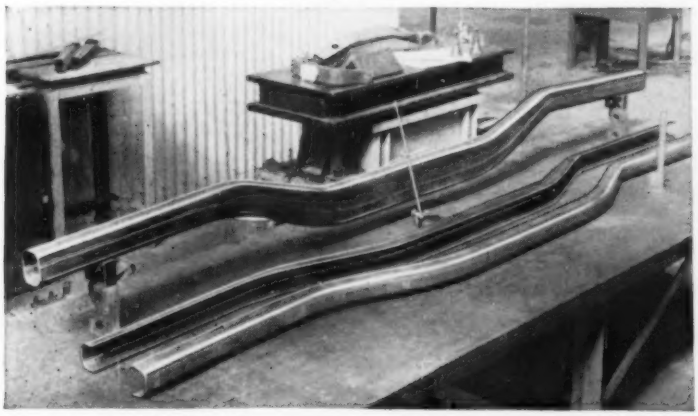

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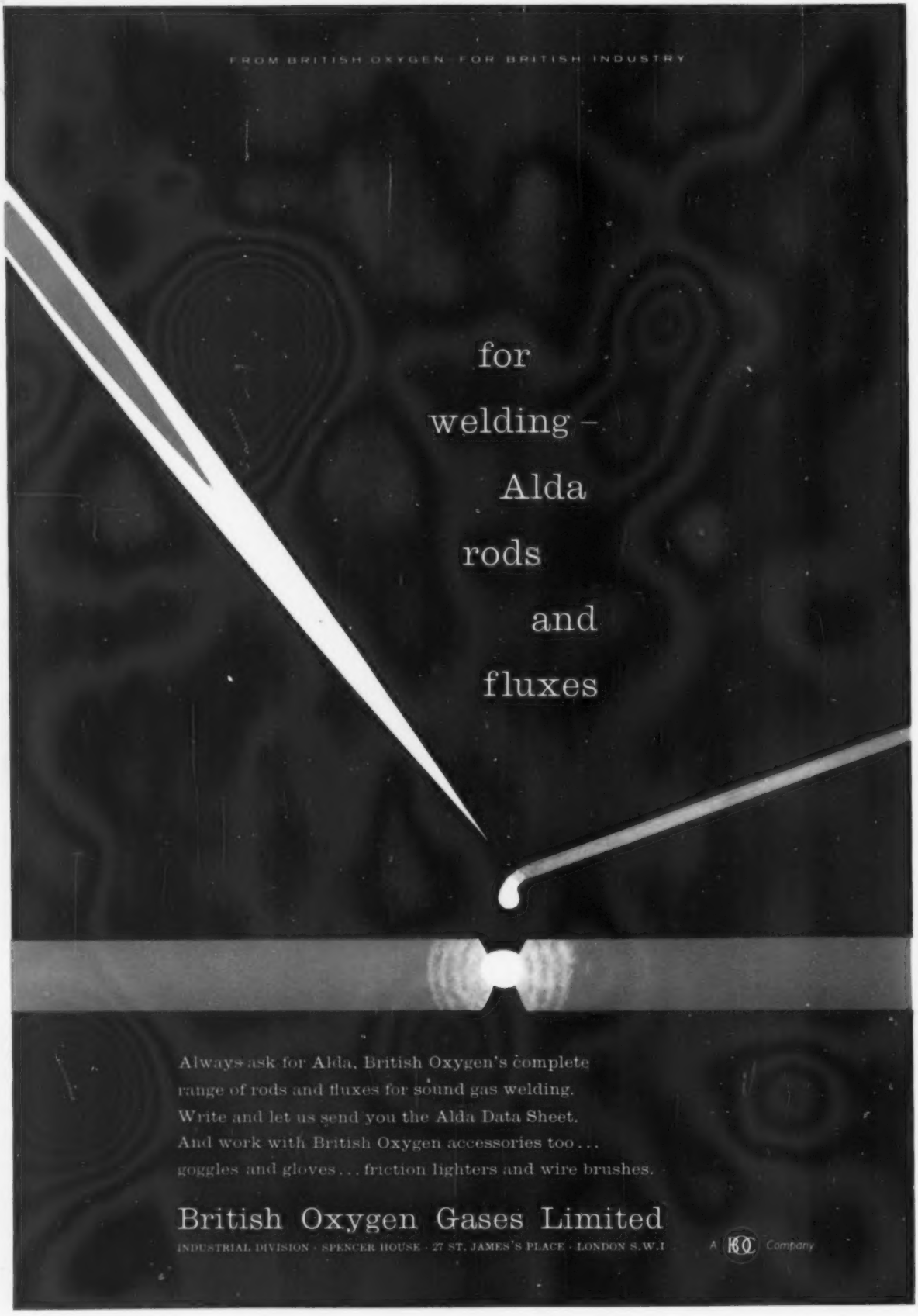
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